Journal of the Royal Society of Arts

NO. 4953 FRIDAY, 10TH JUNE, 1955 VOL. CIII

FORTHCOMING MEETINGS

TUESDAY, 28TH JUNE, at 2.30 p.m. '*The Medal in Art and Society*', by Jean Babelon, Conservateur-en-Chef du Cabinet des Médailles, Bibliothèque Nationale, Paris. Professor Michael Grant, O.B.E., M.A., President, Royal Numismatic Society, will preside. (The paper will be illustrated with lantern slides. Tea will be served afterwards.)

WEDNESDAY, 6TH JULY, at 3 p.m. Annual General Meeting (see Notice below). Tea will be served afterwards.

EXHIBITION OF EUROPEAN MEDALS, 1930–1955

An 'Exhibition of European Medals, 1930–1955' is at present being held in the Library. As was announced in the last issue of the *Journal*, the Exhibition is open on weekdays, including Saturdays, from 10 a.m. to 5.30 p.m. (Wednesdays to 7 p.m.), and will remain open until the 29th June. Admission is free.

A number of tickets are still available for the visits to the Royal Mint on Thursday, 16th June, and there will also be two additional visits on Thursday, 23rd June. Special parties will be formed for these at 10.30 a.m. and 2.30 p.m. on both days. Fellows who wish to take part should apply to the Secretary for tickets (not more than two) as soon as possible.

THE LIBRARY

Fellows are informed that for the duration of the Exhibition of European Medals (from the 9th to the 29th June) the Library will not be available as a reading room or for the service of teas. The Parlour, however, will continue to be in use for these purposes, and books may be borrowed and consulted as usual.

Fellows will also realize that books cannot be posted outside London during the railway strike.

ANNUAL GENERAL MEETING

The Council hereby gives notice that, in accordance with the Bye-Laws, the Two-Hundred-and-First Annual General Meeting, for the purpose of receiving the Council's Report and the Financial Statements for 1954, and for the election of officers, will be held on Wednesday, 6th July, 1955, at 3 p.m., at the Society's House.

(By Order of the Council)

KENNETH WILLIAM LUCKHURST,

Secretary.

VISIT OF THE CHAIRMAN TO PARIS

The Chairman of Council visited Paris last month, to return the visit paid to London by the President of the Société d'Encouragement pour l'Industrie Nationale on the occasion of the Society's Bicentenary. Mr. Munro Runtz handed an address to M. Georges Darrieus, President of the Société d'Encouragement, and this is reproduced below in translation. The original is to be included in the annual report of the French society.

The Address

Monsieur Le President,

As Chairman I bring you greetings from the Council of the Royal Society of Arts, and as Acting President of the Faculty of Royal Designers for Industry I offer you their good will and interest.

My Society was deeply conscious of the honour paid to them during their Bicentenary Year by the presence of your distinguished representative.

I understand that Mr. Chaptal founded your renowned Society upon similar aspirations to those of the Royal Society of Arts and we have therefore a bond of friendship and of purpose.

The Royal Society of Arts is greatly interested in extending its system of bursaries whereby, with the help of industrial establishments, students for art in industry might achieve reward which will enable them to travel in France and other countries, and learn how much advance in art and manufactures is afoot.

I hope that our relationship in coming years may become more intimate and mutually more advantageous.

THE MECHANICAL PROPERTIES OF METALS

Three Cantor Lectures

I. THE TENSILE PROPERTIES OF METALS

by

PROFESSOR HUGH FORD, D.Sc.(Eng.), Ph.D., M.I.Mech.E.

Monday, 14th March, 1955

INTRODUCTION

Metallic materials are subjected to a wide variety of loading conditions under service, and it is not surprising that they react in different ways depending upon these varying circumstances. The tensile properties of metals have, by custom, been determined by the tensile test, but it is important to establish what is meant by the tensile properties, and the extent to which the tensile test does, in fact, assess inherent physical characteristics of the material tested.

The tensile test is a standardized, stylized test which, by long custom, the engineer and metallurgist have learnt to interpret in terms of the behaviour of a material in service under conditions far different from those occurring in the test. But the increasing complexity in design, and the need to use materials more economically under higher working stresses, has led to a realization that the characteristics displayed by materials when subjected to a tensile test are only a manifestation of the basic, inherent mechanical properties. Coupled with the spectacular failures which have occurred lately in a number of engineering structures and particularly in welded ships—failures which on the basis of the traditional tests, such as the tensile test, were quite unexpected and unpredictable—it would seem that, in reviewing the tensile properties of metals, I should review the inherent mechanical properties under general loading, and then attempt to interpret the tensile properties against this general background. It is perhaps appropriate to refer to the remarks of Dr. Desch in the Cantor Lectures in 1940, in which he emphasized that 'the data which the tensile test furnishes are not physical constants', . . . that 'a generalized stress-strain curve should be aimed at', and he questioned whether the time was not ripe for considering this more general treatment of testing as a means of expressing the complete mechanical properties of a material.

THE TENSILE TEST

By tensile properties, we mean those properties which correspond to stresses which tend to pull apart the elements of the material. The tensile test consists

in straining a definite gauge length of the material, and measuring the force necessary to keep the test piece in equilibrium. The usual presentation is in the form of a load-extension graph (Figure 1). There is usually an elastic region in which Hooke's Law holds—that is, the extension is proportional to the load, and if the load is released, the test piece goes back to its original length. At some particular point, this elastic behaviour breaks down, and if the load is released, a permanent extension or set has occurred, that is, the Elastic Limit of the material has been exceeded. The stress at first yield is given by the (yield load)/(original area, A_o), and the value chosen is usually based on the stress at a given off-set, for example, 0·1 per cent permanent set. Further extension leads to an increasing plastic extension, the material generally showing a rising load-extension curve. This shows that the material has an increasing strength with plastic deformation (strain hardening) until a point is reached where the load-extension curve becomes horizontal. Careful measurements have shown that the volume of wrought materials remains constant during plastic deformation within very close limits. Accordingly, the slope of the load-extension curve with increasing extension is the combined result of an increasing yield stress p, and a decreasing cross-sectional area A, such that $p \times A = P$, the current load. So long as the extension Δl is uniform throughout the gauge length, $A \times l = A_o \times l_o$, for constancy of volume. At the point of maximum load, the test length starts to deform locally at the weakest section and the cross-sectional area decreases here more rapidly than elsewhere, so that a neck forms, and the rest of the load-extension curve is unstable, the load falling to maintain equilibrium.

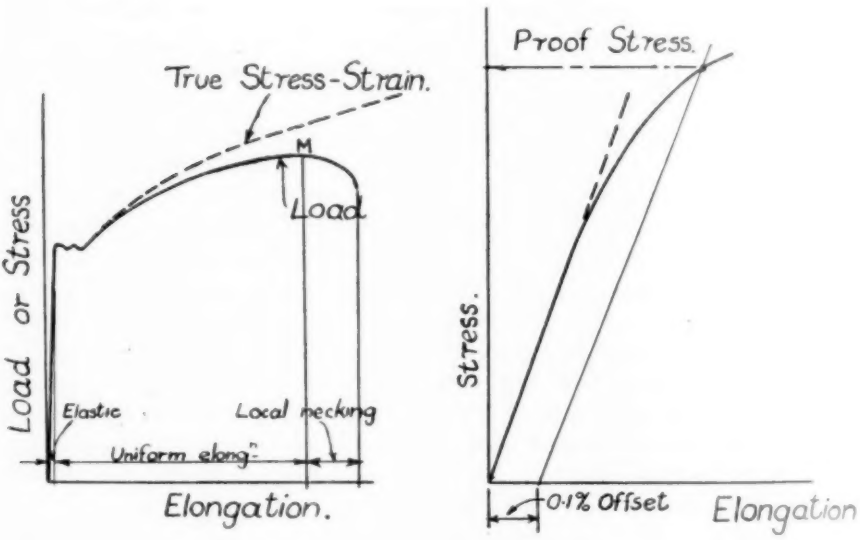

FIGURE 1. *Load extension graphs*

The uniform extension up to the maximum load is proportional to gauge length, while the local extension beyond the maximum load is confined to the neck region and is independent of gauge length. The local extension depends upon the physical dimensions of the cross-section of the test piece however, and in specifying a *total* percentage elongation, it is necessary to state the gauge length. In general testing work, the tensile properties are characterized as follows:

$$\text{Ultimate tensile strength} = \frac{\text{Maximum load}}{\text{Original area}} = \frac{p_m}{A_o}$$

$$\text{Percentage elongation} = \frac{\text{Total extension to fracture} \times 100}{\text{Original gauge length}} = \frac{\Delta l_T}{l_o} \times 100$$

$$\text{Reduction of area, per cent} = \frac{\text{Original area} - \text{final area of neck at fracture}}{\text{Original area}} \times 100 = \frac{A_o - A_n}{A_o} \times 100$$

Some materials have unusual load-extension curves: for example, many steels show a characteristic, sudden yielding without increase in load at first yield, and an extension of as much as four per cent of the gauge length may occur before a state of equilibrium is again obtained, shown by a rise in the curve (Figure 2)†. This effect is now known to occur in certain non-ferrous alloys

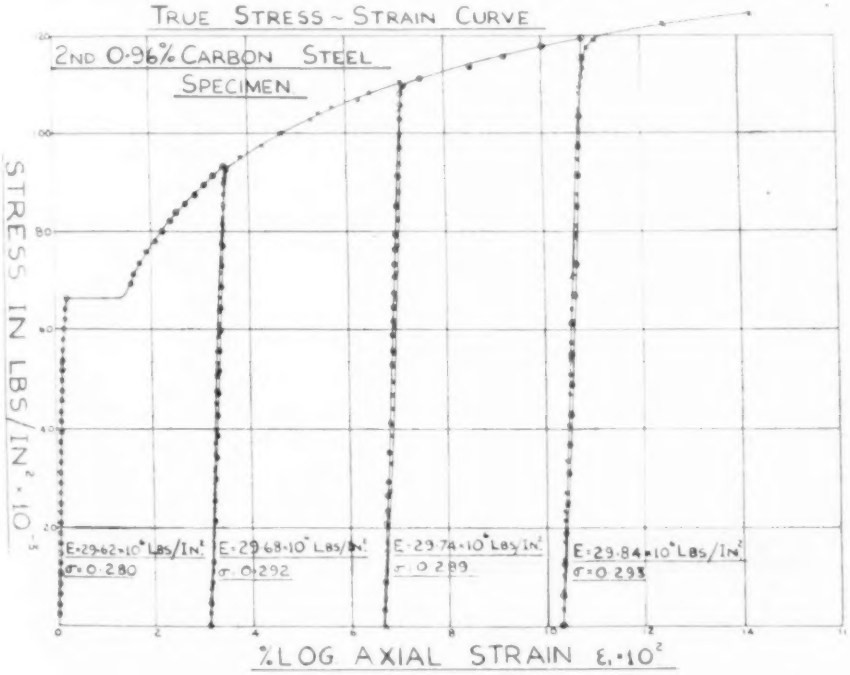

FIGURE 2. *True stress-strain curve*

† The help of my colleague, A. Shelton, is acknowledged in the preparation of these results.

where it may be manifest as a yield point in the mild steel sense, or as a progressively occurring phenomenon (Figure 3), the steps becoming larger as deformation continues[1]. These are ageing effects, which can occur more rapidly with these non-ferrous alloys than with steel, but which also depend upon temperature. The same kind of stepped curve can be obtained with mild steel (Figure 4)† by allowing long rest periods between steps. But generally, the load extension curve is smooth and continuous, although the relative proportions vary considerably and it is necessary to determine to what extent these are real physical characteristics of the material tested.

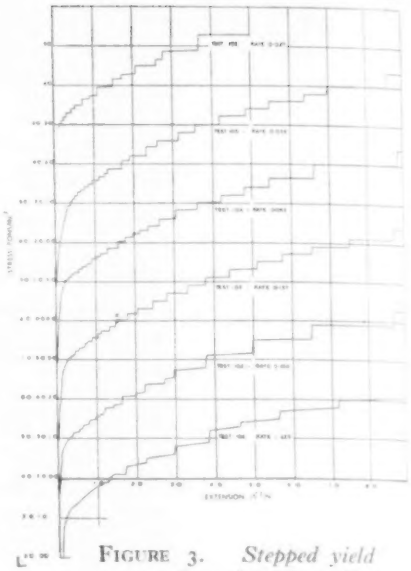

FIGURE 3. *Stepped yield curves of an aluminium alloy*

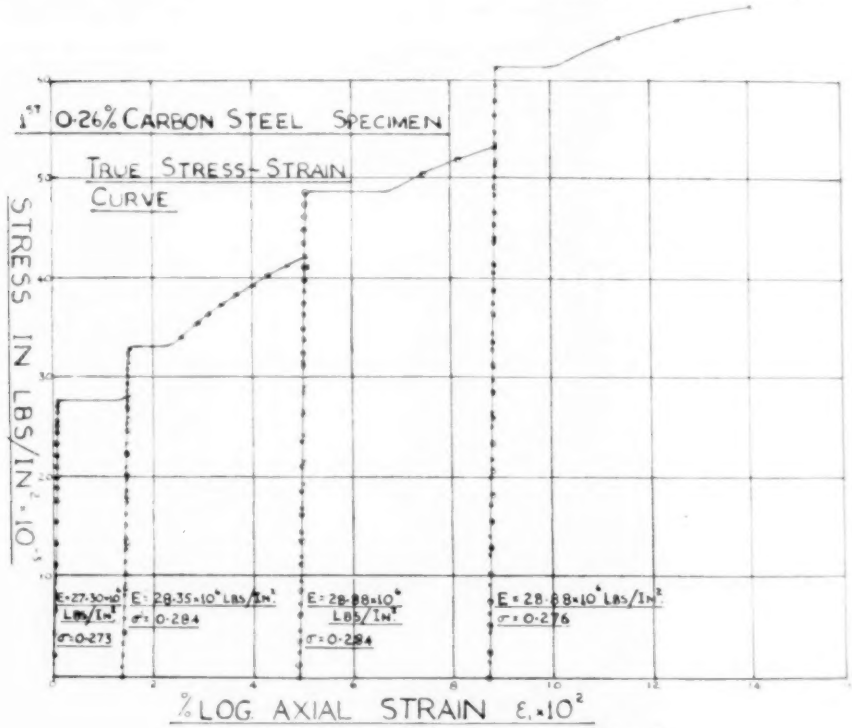

FIGURE 4. *Stepped yield curves for mild steel*

† The help of my colleague, A. Shelton, is acknowledged in the preparation of these results.

The uniform extension which occurs up to the maximum load can with some confidence be related to a real characteristic of the material. Clearly there is a specifiable extension in this region: the extension Δl on length l, being uniform, can be expressed as a *strain* $e_l = \frac{\Delta l}{l_o}$, and the true stress is given by $p = \frac{P}{A}$, where A is the current area and P the corresponding load. Since the volume is taken as constant, the stress can be calculated from the extension without measuring the cross-sectional area directly, giving

$$p = \frac{Pl}{A_o l_o} = \frac{P(1 + e)}{A_o}$$

In this way, up to the maximum load, a curve of stress against strain can be plotted which at least has physical meaning.

Beyond the maximum load, where the neck forms in some region along the length, the deformation is confined to this region, and is not uniform within it. Moreover, the stress is no longer a simple uniaxial tension, because the shoulders of the neck, being less heavily stressed, act as a restraint to the material within the neck, and so exert an outward radial stress. This means that the neck has a complex state of stress, made up of an axial tension, and radial tension (Figure 5).

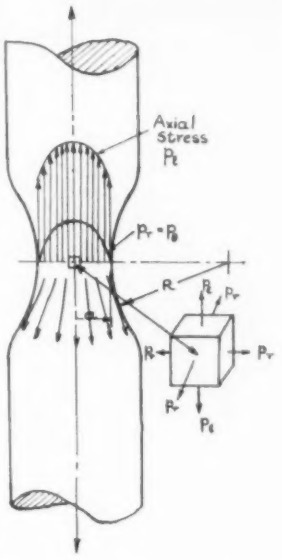

FIGURE 5. *Diagram showing neck in a tensile test piece with variation in stresses across the section owing to constraint caused by the adjacent material*

P_r = *radial stress*
P_θ = *circumferential stress*

Two questions therefore need answering before any attempt can be made to give a physical interpretation of the load-extension diagram beyond maximum load. (1) Why does the neck form? (2) Having formed, how do materials behave under complex states of stress and strain?

COMPLEX STATES OF STRESS

There are two mechanisms of general strength of metallic materials—shear and fracture. On the general conception of crystal structure, these mechanisms represent respectively the sliding of one atomic layer over another, and the tearing of one layer from another by direct forces which exceed the maximum inter-atomic forces. The shear or slip process on the theory of dislocations has been described by Mr. Bailey in last year's lectures and it is only necessary to point out that shear can take place relatively easily because the shearing is looked upon as taking place one atom at a time across the atomic layer.

On the other hand, fracture is looked upon as starting at some point in a cross section, normal to the direction of maximum local tensile stress, where this local stress has a sufficiently high value and where the shear mechanism is prevented from occurring. A crack starts at this point of high tensile stress concentration, by a few atoms being pulled out of the range of each other's

cohesive forces. Once this crack starts to form, it can continue across the section, so long as the local tensile stress is maintained at the requisite value. It is important to notice that the slip mechanism requires a stress tangential to the plane across which the slip is to progress, and that the slip tends to relieve the state of stress; also, after slip, the atoms remain at the same relative distance apart, the only change being that the atoms in one layer have moved over to new neighbours.

A shear stress is, by definition, one which is applied tangentially to the plane on which it acts. But it can easily be shown that a shear stress arises when the external loading system gives rise to differences between the direct stresses (tensile and compressive) on mutually perpendicular planes. For example, a simple uniaxial tension causes a shear stress which has a maximum value of $\frac{1}{2} \times$ (the numerical value of the tensile stress) on a plane at 45° to the tensile stress. An equal compressive stress causes an equal shear stress, but of opposite sign. That is, so far as the shearing mechanism is concerned, the sign of the direct stresses does not matter,—it is only the intensity of stress difference, that is to say, of the shear stress, which decide whether shearing will occur.

This is not the place to go into the complexities of stress analysis, but it can be shown that any loading system, however complex, can be resolved into a state of stress, at each point throughout a continuous medium, composed of three direct stresses (called principal stresses) on three mutually perpendicular planes. No shear stresses exist on these planes, but the critical shear stresses occur on planes at 45° to these principal planes. If we denote the three principal stresses by p_1, p_2 and p_3 then the limiting shear stresses are:

$$\frac{(p_1 - p_2)}{2}, \qquad \frac{(p_2 - p_3)}{2}, \qquad \frac{(p_3 - p_1)}{2}$$

and these are the stresses which, when metals are subjected to complex states of stress, determine the point when yield will occur. The critical condition in terms of these differences for plastic yield to occur is now usually represented by the equation:

$$(p_1 - p_2)^2 + (p_2 - p_3)^2 + (p_3 - p_1)^2 = 2f^2$$

where the value of f must be determined from the physical properties of the material. Figure 6 shows a small cubical element of material subjected to principal

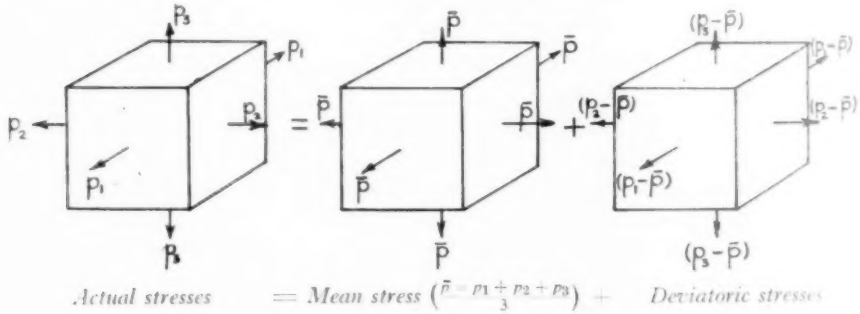

FIGURE 6. *Hydrostatic and deviatoric stress components*

stresses, which can be broken down into two equivalent stress systems, a hydrostatic component $\bar{p}$ = average stress (that is, a stress uniform in all directions), and the deviations from this hydrostatic mean stress $p_1' = p_1 - \bar{p}$ and so on. The mean stress can have a large value, but so long as the *differences* between the principal stresses do not exceed the critical condition for the onset of plastic distortion, the intensity of the mean stress has no effect on yield.

The effect of a direct compressive stress is to force the atoms closer together, so that a 'hydrostatic' compression produces an elastic volume change which is recovered on releasing the stress. A uniform tensile stress has the same effect, but of opposite sign, as a compressive stress, in that recoverable elastic change of volume results; but in the limit, a 'hydrostatic' tensile stress can have the additional result of causing a crack to start when the cohesive strength is reached at some point of weakness.

Summarizing, there are three ways in which materials can behave under the application of external forces: they can deform elastically, they can deform plastically, or they can break. In the elastic range, both direct and shear stresses cause corresponding, but very small, distortions, the atoms moving nearer or further apart, but always returning to the same configuration on removal of the stresses. Beyond a certain critical state of shear stresses, permanent deformations occur. It is clear from this discussion that permanent deformation, since it depends upon the shear mechanism, takes place without change of volume, and since relatively large displacements occur under high stresses, such deformations are accompanied by considerable absorption of energy. On the other hand, fracture by direct tensile stress requires only that the atomic layers should be torn apart by a very small distance to decrease the cohesive force to negligible proportions, so that fracture occurs with little energy absorption.

This qualitative picture is perhaps a little over-simplified, but it will be clear that what is known as a ductile material has a shear strength, that is, a resistance, to general shear, which is lower than the cohesive strength, that is to say, the resistance to rupture and, conversely, a material is brittle if it has a high shear strength, and the material reaches its rupture limiting stress before it can shear.

DEVELOPMENT OF THE BASIC YIELD STRESS CURVE

It has been usual to classify materials as brittle or ductile but these terms are relative. Several factors such as temperature, metallurgical factors, impurities and heat treatment affect the relative resistance of metallic materials to shear or fracture. However, in discussing the tensile properties of metals, it is the effect of the imposed system of stresses with which we are mainly concerned. Many years ago von Karman[2] showed that marble, which is a brittle material, shattering under simple uniaxial stress after imperceptible plastic strain, could be made to behave in a distinctly ductile manner merely by superimposing a high hydrostatic stress. These sort of experiments have been repeated and extended by Bridgman.[3] On the other hand, by notching a tensile testpiece of mild steel, so that high hydrostatic tension is produced, an otherwise ductile

material can be made to fracture with negligible deformation and absorption of energy.

How then, are the intrinsic properties of materials to be assessed, since from these considerations they seem to be relative to the test conditions? Can we increase the ductility of normally ductile materials, and is there any basic yield stress-strain curve for a given metal from which its behaviour under any particular testing conditions can be predicted? Let us examine recent experimental work, to see what has been done in this respect.

Simple Compression

Compression tests have often been proposed as a means of obtaining a stress-strain curve, but there are a number of difficulties which present themselves, mainly owing to friction between the compression platens and the ends of the test piece. This gives rise to 'barrelling' and, very rapidly, simple uniaxial compression no longer exists, so that the measured load cannot be expressed as a stress, neither is there a constant cross-sectional area to which it could be referred. It is found, however, that by using highly polished platens, and certain special lubricants, and if, in addition, the load is applied in small increments, releasing the load to allow re-lubrication between each increment, the friction remains at a very low value. It seems that so long as the relative movement—that is, the deformation—between load increments is kept small friction has a very small effect, as judged by the amount of barrelling. This technique has the advantage that, since the testpiece is removed from the testing machine between each increment, the amount of compression can be accurately measured.

Further, by using four testpieces of the same material, of identical diameter but of different lengths, an even closer approach to the ideal case of zero friction can be obtained.[4] As the ratio of height to diameter increases, the constricting effect of friction at the ends becomes a progressively smaller contribution to the deformation, and at 'infinite' height the friction effect would theoretically vanish. Thus by loading all four testpieces incrementally, each in turn to the same series of loads, the amount of deformation produced is plotted against the ratio of $\frac{\text{diameter}}{\text{height}}$ for each test length and extrapolated to zero—that is to say to infinite height. This gives us the deformation of an ideal test without friction, and the true stress is given by dividing the load by the cross-sectional area, calculated from the percentage deformation figure. This method is fairly time-consuming, but is accurate and needs no special apparatus, and the compression test has the outstanding advantages that it is inherently stable (unlike the tension test, because the cross-sectional area increases with deformation) and much larger deformations can be obtained. That is, the stress-strain curve can be extended much further than in a tensile test on the same material.

Plane Compression

A second method makes use of incremental loading in compression (this was, in fact, the method for which it was developed) but on strip material between

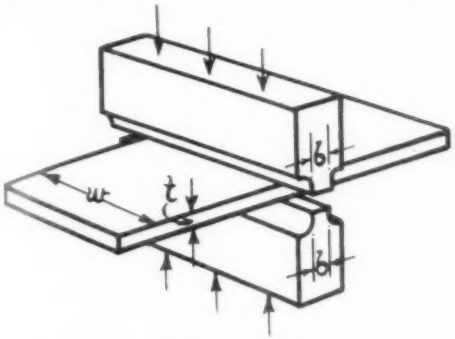

FIGURE 7. *Deformation in compression*

dies[5] (Figure 7). The non-deforming material on each side of the dies acts as a restraint to lateral spread of the strip, all the deformation going into elongating the material between the dies.[5] The deformation pattern is somewhat complex, but as the ratio (die breadth)/(strip thickness) increases beyond 2:1 it approaches closely to deformation on two planes at 45°, in which case the yield stress is exactly $\sqrt{\frac{2}{3}} = (1{\cdot}155) \times$ the yield stress in simple tension or compression. Naturally, we cannot increase the $\frac{\text{breadth}}{\text{thickness}}$ ratio too far because friction again intrudes, even with incremental loading and lubrication, but it is found that if the ratio is kept between 2:1 and 4:1 the deformation is very closely maintained to plane compression on 45° shear lines. This method has distinct advantages over simple compression, owing to the effect of the overhanging material on each side, which forces the plastic zone to deform according to a certain flow pattern, rather than to the frictional restraint.

Figure 8 shows curves for high conductivity copper obtained in this way, using various thicknesses of strip, but always keeping the ratio of $\frac{\text{die breadth}}{\text{strip thickness}}$ between 2 and 4.[6] This basic plane strain curve should—if there is a basic yield stress curve for a metal—be capable of relation to the simple compression curve: the stress ordinate should merely differ by the factor 1·155, while the

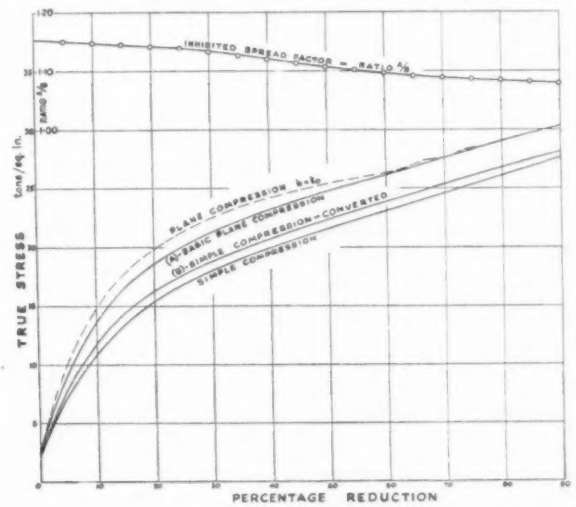

FIGURE 8. *Yield stress curves for high-conductivity copper*

strains occurring in the two cases, while they will be different, can be exactly specified—by means of a mathematical relationship between the strain quantities —in terms of the reduction in height. Thus, in Figure 8, the lowest curve is simple compression, curve B simple compression corrected to the same strain basis as A, the basic plane compression curve. At the top, the ratio of the true stress values A/B is computed for various reductions. It is a remarkable fact that this curve cuts the ordinate for zero reduction at 1·155, but falls steadily with increasing deformation to a value of about 1·08 at ninety per cent compression.

Herein lies one of the main difficulties in trying to obtain a basic yield stress curve; although by changing over to compression we can extend the total uniform strain about twenty times compared with that in the tensile test, another factor enters in that metallic materials develop directional properties. For example, the yield stress in the transverse direction (that is, in the direction in which deformation has been prevented) is about 8 to 13 per cent. higher than in the longitudinal direction in the plane strain test, and the anisotropy will develop differently under different strain systems.

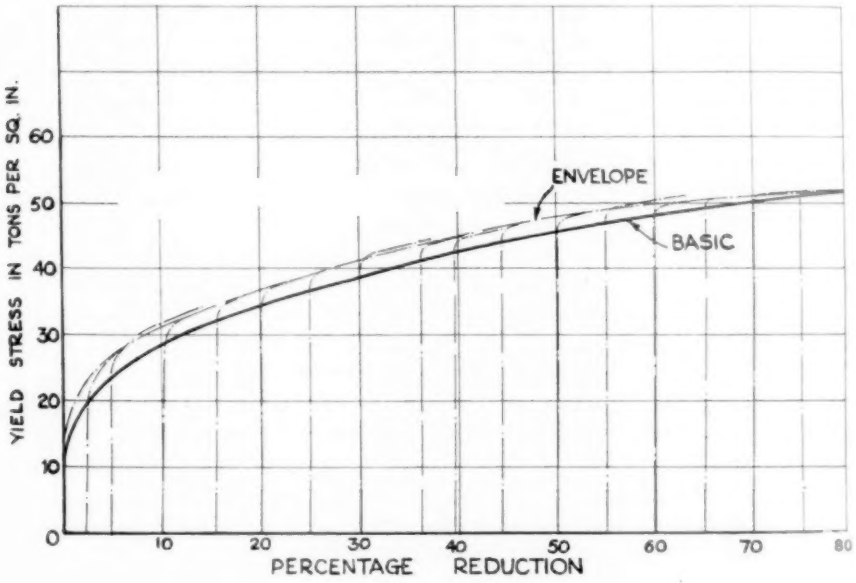

FIGURE 9. *Yield stress curve for mild steel*

Plane strain compression is closely approximated in cold rolling, and Figure 9 shows curves obtained for mild steel by the plane strain test, both in strip compressed entirely between the parallel dies in the test, and also on progressively rolled strip, similarly tested after each pass.[7] The difference between the curves is the additional plastic straining in cold rolling owing to the curvature of the rolls which causes an additional shearing strain.

It may be asked whether any means exist for extending the tensile yield stress curve beyond the point of necking by an essentially tensile stress. This can be done in at least two ways. First, by wire drawing. In this process, the wire is pulled through a series of dies of decreasing throat diameter in a succession of passes, and overall elongations of several hundred per cent can be achieved without the wire breaking. Here, the tensile force needed to pull the wire through the die is supplied by the wire itself on the downstream side of the die. The compressive stress supplied by the die wall is sufficient to cancel the radial tension which arises as soon as the neck forms in the tension test, and although this stress has a component opposing the tension in the drawn wire (as has also the frictional force to which it gives rise) nevertheless, the net state of stress at each point is such that yielding can occur more easily in the die throat than elsewhere, and the wire does not neck on the downstream side of the die, if the pass reductions are suitably chosen. This does not mean, of course, that the wire cannot be broken by deliberately attempting too big a reduction in one pass, but only that by a predominantly tensile stress, very large strains can be obtained. As in cold rolling, there is a non-uniform distribution of stress and strain in the die, so that a test carried out on a piece of the wire after each pass does not give a basic stress-strain curve, but this process serves to illustrate that elongations of several hundred per cent are thus possible, compared with about forty per cent for the most ductile material in simple tension, merely by suppressing the radial tensile component of stress.

An even more striking example is provided by the hydraulic bulge test.[8] In this test, a circular disc of the test material is clamped around its edge, and hydraulic pressure is applied to one side, causing it to stretch and bulge into a roughly spherical form. The total plastic strain up to the point of instability, that is, the maximum load point, at which the neck would start in simple tension, is computed from measuring the thinning at the pole, and with most metallic materials this total strain is at least twice that in simple tension on the same materials. This is a stress system made up of biaxial tension—that is to say, in the plane of the sheet there are equal tensile stresses at right angles, with a negligible stress in the thickness direction.

INTERPRETATION OF THE TENSILE TEST

In the light of all this evidence, and much more which could be put forward, we must now consider what is meant by the tensile properties of metals, and how we are to interpret the results of the tensile test.

First it is clear that where the test piece is free to shear, or where the maximum tensile stresses are forced to remain below the cohesive strength, then metallic materials show very considerable capacity for deformation—that is, they are very ductile. The stress differences alone decide whether shearing will occur, and, so far as yield is concerned, tensile and compressive stresses have the same effect. This is seen from the form of the yield criterion

$$(p_1 - p_2)^2 + (p_2 - p_3)^2 + (p_3 - p_1)^2 = 2f^2$$

since all the stress differences are squared, their sign is always positive, irrespective

of whether the stress difference is positive or negative. The value of f depends upon the material and the degree of work-hardening and, in principle, is taken off the basic stress—strain curve obtained from one or other of the methods already described, or from the torsion test.

The properties displayed in any test depend entirely upon the state of stress to which the test subjects the material and while ideally, by ensuring that the material can shear, a basic stress-strain curve can be developed up to very large strains (which are capable of being converted to the strains corresponding to any other test condition), nevertheless, with technically available metals, anisotropy and other effects intrude.

Where, however, the state of stress involves predominantly tensile stresses, the second mechanism, that of tensile rupture, can interfere with the shear mechanism. Rupture is really the characteristic tensile property.

Consider, then, a circular tensile testpiece, which is pulled slowly until it finally fractures. It has been shown that up to the maximum load, the extension being uniform, the yield stress-strain curve is accurately calculable and definable, except, of course, for the possible development of anisotropy. There is one essential difference between simple tension and simple compression: the compression test is inherently stable, while tension is not: with continued plastic strain, the yield stress increases, but the rate of hardening slows up with increasing strain, and the area supporting the internal load decreases. When the curve reaches the point where the proportional decrement of cross-sectional area exceeds the corresponding proportional increment in yield stress, instability sets in, and this point can be shown to correspond to the maximum load, that is,

$$\frac{1}{p}\frac{dp}{de} = \frac{-1}{A}\frac{dA}{de} \text{ *}$$

The extent of the general elongation up to maximum load depends therefore upon the strain-hardening capacity of the material. The neck forms because of this. The greater the slope of the stress-strain curve the greater is $\frac{1}{p}\frac{dp}{de}$, and the greater can the uniform extension be (and correspondingly the smaller the area A) before this equality is established. This leads to the conclusion that the uniform

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* This can be shown thus :

The strain hardening rate is $\frac{dp}{de} = \frac{d\,P/A}{de}$

$A = \frac{A_0}{1+e}$ so $\frac{dp}{de} = \frac{1}{A}\frac{dP}{de} + \frac{P}{A_0}\cdot d\frac{(1+e)}{de} = \frac{(1+e)}{A_0}\frac{dP}{de} + \frac{P}{A_0}$

When P reaches its maximum $\frac{dP}{de} = 0$ and $\frac{dp}{de} = \frac{P}{A_0} = \frac{P}{A}\cdot\frac{A}{A_0} = \frac{p}{(1+e)}$

So the proportional increase in strain hardening is $\frac{1}{p}\frac{dp}{de} = \frac{1}{1+e}$

The proportional increase in area is $\frac{1}{A}\frac{dA}{de}$

$\frac{1}{A}\frac{dA}{de} = \frac{1}{A}\left(d\frac{A_0}{1+e}\right) = \frac{-A_0}{A}\cdot\frac{1}{(1+e)^2} = \frac{-1}{1+e}$

∴ the maximum load is reached when $\frac{1}{p}\frac{dp}{de} = \frac{-1}{A}\frac{dA}{de}$

extension is of greater value in assessing material properties than the total elongation, both from the point of view of real ductility, and as a means of deciding the reservoir of strength represented by the increase in yield stress between first yield and maximum load.

We must now consider the load-extension curve beyond the point of maximum load. We have to suppose that one section of the testpiece is weaker than all others, and at the maximum load this section will have a smaller value of

$$\frac{-1}{A}\frac{dA}{de}$$

so that plastic flow occurs more rapidly here than elsewhere, leading to the formation of a neck. This is not inconsistent with the observation that sometimes two necks start to form, although only one finally develops.

When we pull a tensile testpiece in a testing machine, plastic deformation of the testpiece occurs, but it also suffers elastic extension. The former is localized at the neck, but to the latter all parts of the testpiece contribute. Moreover, in most cases, the elastic strain energy of the testing machine is in part available for rapid release, as soon as the unstable point is reached. With a fairly rigid testing machine, the elastic line of the machine plus that part of the testpiece excluding the neck, has a high spring constant, say C. Thus, if the neck extends plastically by distance dx, the load on the testpiece must drop off by an amount dF by the elastic equation

$$dF = -C\,dx$$

So long as the slope C is greater than the slope of the load extension curve, in the unstable region, the load will drop off more rapidly than the force required for further plastic yielding, and load-extension measurements are possible at this stage. When the slope of the plastic curve and elastic constant of the system are equal, further extension occurs almost instantaneously and no measurements can be made beyond this stage. The test 'runs away', and the excess energy goes into the noise of the fracture.

Reference was made earlier to the stress system which develops at the neck, owing to the restraint of the less heavily stressed material on each side. The stress system at the minimum section can be looked upon as made up of two parts, the uniaxial tension, which by giving rise to a shear stress determines yielding, and a hydrostatic tension, whose value depends upon the radius of curvature of the neck at its minimum section, and varies from zero at the outside surface rising to a maximum at the central axis (Figure 5). Bridgman and others have shown this picture to be the correct one, and it suggests at once why the fracture of a circular testpiece so frequently has the characteristic cup and cone appearance. The material at the bottom of the neck has been very heavily strained and therefore has a high shear strength. On the other hand, there is no evidence to suggest that the cohesive strength of materials is much increased by strain and, as the neck continues to contract, the triaxial tension stress progressively increases until the sum of this stress and the uniform axial stress component together reach the cohesive strength, and a small crack starts at the axis, spreading out radially

until the triaxial tension has been relieved. Shearing can then again occur in the outer ring of the neck, which therefore tears apart on a 45° surface with a typical shear type of fracture.

The question arises as to whether it is possible to extend the true yield stress-strain curve, making use of our knowledge of the stress and strain distribution in the neck. MacGregor[9] has in fact proposed to measure the diameter—and hence the area of the minimum section,—and to calculate the true strain from the simple fact that $A_o l_o = Al$, so that

$$\Delta l = \frac{l - l_o}{l_o} = \frac{A_o - A_n}{A_n}$$

where A_n = area of the neck. For a very small length at the base of the neck the minimum diameter can be considered constant, and so long as the strain is calculated on this section, this equation will give the true strain e_1 until actual fracture starts.

Bridgman[3] shows that the corresponding axial tensile stress is given by an expression of the form

$$p_1 = \frac{P}{A_n \left(1 + 2\frac{R}{a}\right) \log \left(1 + \frac{1}{2}\frac{a}{R}\right)}$$

where R is the radius of curvature of the neck and a is the neck radius. In principle, therefore, it should be possible to extend the yield stress curve in tension to a point just before fracture, but so far no detailed comparison has been made with the yield stress curve obtained in this way and by the other ways I have described. It is to be expected that the anisotropy would be different in the different methods of loading, but that otherwise the variations should be negligible.

We can now summarize the stage we have reached:

(1) The mechanical properties displayed by a material depend upon the conditions of test, but these apparent properties are a reflection of the basic properties.

(2) There is a general basic yield stress—plastic strain curve, but metallic materials do not remain isotropic, and they suffer from time effects, such as strain ageing, so that a basic curve free from these modifying effects is difficult to obtain. Thus the system of loading affects the stress-strain curve, even where the imposed stress system remains unchanged throughout the test.

(3) This stress-strain curve is a consequence of the shearing mechanism of metals and this is the mechanism of permanent plastic deformation. Where the conditions of loading are predominantly tensile, or the shear mechanism is prevented, a second mechanism—that of tensile rupture—may become operative.

(4) The tensile properties of a metal, as indicated by the tensile test, are a combination of these two mechanisms, but some plastic deformation

however small, seems to be necessary before the rupture mechanism starts.

(5) So far as plastic deformation is concerned, the *tensile* properties of metals are no different from those of any other form of stress, stress difference being the controlling factor, both in causing yield and in maintaining it during subsequent plastic deformation.

(6) But tensile stresses are a necessary condition for rupture and crack propagation, the incidence of rupture being dependent upon the suppression or augmentation of the absolute value of tensile stress.

(7) Plastic deformation increases the shear resistance of materials, but does not appear sensibly to change the rupture strength. Thus the relative intensity of shear to absolute tensile stress to which the material must be subjected, changes with the amount of cold work. Therefore, a cold worked material is more susceptible to fracture because its shear resistance is higher relative to its rupture strength than in the case of a non-work-hardened one.

(8) Mechanical tests on materials have three main functions:

(*a*) To measure the basic properties of materials.

(*b*) To form a basis for predicting the behaviour of materials in service.

(*c*) As a recognized standard for inspection and acceptance tests.

(9) The quantities, such as ultimate tensile strength, percentage elongation, reduction of area, and so on, have no physical meaning in terms of the basic properties of materials.

BRITTLE FRACTURE IN MILD STEEL

No lecture on the tensile properties of metals would be complete without reference to the spectacular failures which have occurred in welded ships. The American Liberty and Victory all-welded merchant ships of the war years first showed that mild steel was behaving in a way which could not have been predicted by the tensile test, and while it is true to say that in the early cases poor welding and design were somewhat to blame, it soon became clear that the trouble was not cured by improving the structural design or paying closer attention to welding.

The essence of the problem is that all mild steel in the presence of a notch or stress concentration fails under tensile stress in a brittle manner—that is, in a sudden extensive manner and with negligible plastic deformation or absorption of energy—when the temperature is lowered sufficiently. This type of failure is particularly sensitive to impact loading. There is then a critical temperature range for every mild steel, above which brittle fracture at a notch does not occur, and below which it does occur, with a narrow transition range in between. The stress needed to propagate the crack is usually lower than that needed to initiate it.

It would seem that the shear strength of metals rises rapidly with decreasing temperature while the rupture strength does not; the presence of a notch causes a concentration of tensile stress, just as the neck does in the tensile test, and

under these circumstances, as we have seen, failure by tensile fracture can occur, a crack opening up from the apex of the notch.

The catastrophic nature of these failures in ships is emphasized when the ship's hull is welded throughout. In a rivetted ship, if a crack started in one particular plate, it might run across that plate, but would stop when it came to the opposite edge. In a welded ship the plate is continuous and the crack is not arrested in this way. Moreover, more of the elastic energy stored in the ship's hull is available to maintain the running crack than there would be in one or two plates only in the rivetted structure. The further the crack runs, the more energy is released to feed the crack, so that it propagates at very high speeds. This explains why the ship gives so little warning of disaster.

Brittle fracture in mild steel structures is a very complicated problem, with many contributory factors, but the important point to emphasize here is that the normal test procedures, and in particular the tensile test, did not allow this behaviour to be predicted.

Some alternative form of test must therefore be devised, and this serves to conclude this lecture: that the data given by a test are as much a characteristic of the test as of the material. If we are to be successful in the best use of materials, we must relate their inherent properties to service conditions through critical test procedures.

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II. CREEP

by

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I must begin with an explanation and, in some sense, with an apology. Where the question of creep in metals is concerned the practical man, who rightly is in control to-day both of money and policy, is interested in very small and slow systematic movement, over periods of thousands if not hundreds of thousands of hours, of complicated alloys. In an hour, however, it is only possible to deal with generalities of some comprehensive importance, and in the field just indicated there are, as far as I know, no generally valid regularities, let alone theories. Further, even if I had wished to talk of a selected practical problem, or problems, of creep-resistant alloys, I am known not to be qualified to do so. I shall mainly concern myself, then, with simple metals and the evidence which they offer as to the mechanism of creep. The work with which I shall mostly be concerned is open to the criticism that the duration of the tests concerned is often—but not always—a matter of hours rather than tens of thousands of hours and this is a valid criticism. What the practical man wants is a short-time test which shall enable him to foretell behaviour during extensive periods of stressing. That we cannot supply, but investigation of short time behaviour may be a beginning, an approach from one side that may meet an approach from the other side and lead to a tunnel through the mountains of accumulated and rocky obstruction—and abstraction. For when theories cannot be tested by experiment their interest is mainly aesthetic.

I defend the academic approach, then, on the ground that it may lead to an understanding of the physical factors involved in creep, and that such an understanding is necessary if we are ultimately to design short-time tests that will give us information about the behaviour over extended time. Any such test will, of course, have to be carefully checked by long-time experiments in the first case, but the time that will ultimately be saved when it has proved itself will justify the preliminary labour. An understanding of the factors involved must, further, prove the best basis for research on new alloys. I hope to show you that a beginning has been made towards a knowledge of some of these factors and if you reply that it is a small and scanty beginning I shall agree with you, but point out that a study of why a red-hot poker, or a glowing platinum wire, discharged an electroscope might seem a small beginning towards the glory of the age, television, but it was the small beginning and it was academic.

If a definition is required, I define creep as the slow progressive change of strain that takes place under stress. Perhaps it would be well to begin with a word on methods of measurement. The great majority of creep measurements have been made on the extension of cylindrical wires or rods under tension, very ingenious methods having been employed in some cases, notably at the National

Physical Laboratory, for the accurate measurement of the movement. In such tests, as the specimen extends, the cross section diminishes and so the stress increases. I have always contended that results are likely to be easier to interpret if some device is adopted which systematically decreases the load during the test so as to keep the stress constant. This point of view has won adherents, but it is still held that constant load is good enough if the greatest extension to be measured is only one, or possibly two, per cent. However, in certain cases, an increase of stress of one per cent, as against constant stress, leads to an increase of extension of some 12 per cent. Before deciding to work at constant lead it is well to examine roughly how the behaviour of the metal in question is varying with stress at the temperature, and in the region, in question.

The devices for maintaining constant stress have often been described. I will only refer in passing to the Andrade-Chalmers bar which, with variation, has been extensively used. A completely different method of measurement is that where the metal is subjected to simple shear, but that is, perhaps, better considered a little later.

The general result of these early experiments was that the results for the metals in question—lead, copper, iron, tin and mercury, to which cadmium has since been added—could be expressed by

$$l = l_0 (1 + \beta t^{\frac{1}{3}}) e^{\kappa t} \qquad (1)$$

The exponential is merely a consequence of the increase of length during flow: if $\beta = 0$, then $\frac{1}{l}\frac{dl}{dt} = \kappa$. If κt is small, or shear is in question, we may write

$$\text{strain} = \gamma_0 + \beta t^{\frac{1}{3}} + \kappa t$$

which is the form used by Cottrell and Aytekin to express the flow of single crystals of zinc.

Equation (1) is valid over a wide range of temperature and expresses the fact that, when temperature is taken into consideration, the general behaviour of all pure metals is the same. Roughly speaking, the type of creep curve that prevails, that is to say, the relative importance of β and κ, is given by the ratio θ of the absolute temperature of the test to the absolute melting point: $\theta = \frac{T}{T_m}$. For instance, iron at 450°C. ($\theta = .40$) gives similar curves to lead at 0°C. ($\theta = .45$) and copper at 15° ($\theta = .21$) to lead at −180°C. ($\theta = .16$).

For any consideration of the nature of creep it is important to remember that creep behaviour is not expressible by an equation of state, or, in other words, that the stress and the temperature do not determine the behaviour of a given specimen, even when supplemented by a partial history of past events. Generally speaking, a full mechanical history from a standard state is necessary, even when a metal of given purity, grain size and grain orientation is concerned. Two examples, which are of great importance, may be given. The first is that prestrain, any strain imposed before the test, has a great influence on subsequent behaviour. In my original papers I showed that rapid preliminary extension or torsion diminished and, if sufficiently severe, completely eliminated the β flow in subsequent test. I do not know that anything further has been done on pre-torsion. Kennedy has of recent times investigated the effect of rapid pre-extensions of different magnitude and applied at different temperatures,

with interesting results. In general it may be said that in this way the character of the subsequent flow can be modified as much as desired in the direction of pure κ flow, which can be attained with suitable rapid pretreatment. The slight acceleration of flow with time which takes place during test after very severe pretreatment is no doubt due to recovery.

The other example is the effect of a rest-period during creep tests, which leads to recovery. When the β-flow is in question, reloading at the original load after a period of no stress leads to creep at a rate in excess of that before the interval, which gradually diminishes until, after a sufficient time, the creep behaviour approximates to what it would have been if no interruptions had been made. The recovery during the unloading period depends upon the time of rest, the temperature and other factors. What has not been determined is whether the recovery when the wire is kept stressed during a period of no change in strain (which can be done by so diminishing the stress with time that the creep is exceedingly small throughout the period in question) is the same as when the wire is completely relieved of stress. I had a scheme for carrying out such experiments.

The general lesson is that to get consistent creep behaviour one must start with a standard metal, not only in composition but also in internal state and in grain size. Lead anneals itself at room temperature in course of time, but it is better to anneal it systematically: most other metals need systematic normalization if they are to give consistent results. Aluminium, which seems to be almost the only metal on which creep research is carried out to-day, is always carefully normalized.

A general result of the early measurements was that creep can be considered as consisting of two parts which, if not entirely independent, yet express two different mechanisms and have a physical reality. The third stage of accelerating creep I will consider later. The $t^{\frac{1}{3}}$ law has a wide validity: it has been verified on a variety of substances, such as plastics and mortar, as well as on many metals, by workers both in my laboratory and in other institutions, for example, by Dr. Hopkin working on lead over periods of hundreds of hours at the British Non-Ferrous Metals Research Association. Frequently failure to make the formula fit experimental results is due to a wrong choice of constants β and κ. I will quote two examples. Hasaguti claimed that his four-constant formula gave a better fit to the short-time creep of commercial aluminium than the three-constant formula (1), but with a suitable choice of constants this latter formula gave the better fit, and a very close one. Again, Latin did not find a very good fit to certain of his experiments with $t^{\frac{1}{3}}$, but Professor Nabarro and I, independently, showed that here again, with the right constants, excellent agreement was obtained.

It is not claimed that $t^{\frac{1}{3}}$ is of universal validity or that it is an exact expression of the process which β is taken to measure, which, as will be pointed out, is very complicated. What is suggested is that it gives a close representation of experimental results in a wide range of cases—as, indeed, work in a variety of laboratories has shown—and that it is often the most convenient method of separating two real and significant processes, for the existence of which there is a great body of experimental evidence.

Wyatt has recently made a careful examination of the transient creep of copper and aluminium at temperatures from −196 to 140°C. with durations up to four days and with maximum strain approaching one per cent, and found that at higher temperatures and stresses the $t^{\frac{1}{3}}$ law held well, but that at low temperature and stress a law in which the strain was a linear function of log t gave a better accord with the observations, although, judging by the curve given, the representation was not very close except when the final strain was not greater than 0·3 per cent. Other workers have varied the exponent slightly from $\frac{1}{3}$. But, on the whole, the $t^{\frac{1}{3}}$ law has proved serviceable, especially where permanent flow is taking place at the same time.

The typical variation of β with stress and temperature is shown in my old papers. A few years ago I found that, for one metal, all the β-against-stress curves could be superposed by multiplying the stress scale by a very simple function of T, or $\beta = f(n\tau)$

$$\text{where } \tfrac{1}{n} = g\left(\tfrac{1}{T} - \tfrac{1}{T_F}\right)$$

Here τ is the stress, T the absolute temperature and T_F is a critical temperature which, for ordinary metals which do not recrystallise under stress, is the melting point. The β–against-stress curves for all metals tested, at various temperatures, can be reduced to one curve by change of scale. I show you, as examples, the diagram for lead at various temperature and that for copper, iron and cadmium, with some figures for other specimens of lead. I must not pursue this further here, and merely bring it in to emphasize that β is a significant constant.

The general variation of κ with stress is logarithmic, that is, $\log \kappa$ is linear with stress. The variations with temperature can be expressed with the help of the exponential law established for the viscosity of liquids. I shall say a word on this when I come to consider creep theories.

There is a body of evidence to support the view that transient creep is connected with processes occurring within the grain and permanent creep mainly determined by processes taking place at the grain boundaries. We shall return to this question after having considered some of the recent work on details of crystal deformation within the grain, but I want to direct attention to two simple experimental methods that indicate such a separation.

First of all, the behaviour within the grain. Chalmers and I, seeking information on modifications of structure, carried out experiments on the electrical resistivity of the metal of wires during the creep process. Various metals, stresses and temperatures were used, the creep being analysed into its l_0, β and κ components in each case. With the face-centred cubic metals copper and aluminium, there was no change of resistivity at any stage of the extension. With cadmium, at various temperatures from 0° to 100°C., there was a decrease of resistivity as creep proceeded: with tin the resistivity increased. This accords with the supposition that a similar rotation of the operative glide planes takes place in the grains in each case, for with both tin and cadmium the electricity resistivity is greatest along the unique crystal axis, but whereas with hexagonal cadmium this unique axis is normal to the glide plane, with tetragonal tin the unique axis is in the glide plane. With the cubic crystals the resistivity is independent of the direction.

so the negative result with metals of cubic structure agrees with our supposition and also indicates that intergranular disturbance does not appreciably affect the resistivity.

Extensive experiments were carried out with cadmium, which showed that the change of resistivity was closely associated with β flow. Immediate extension, amounting in some cases to more than seven per cent, led to no change of resistivity, neither did β flow alone. The experimental change of resistivity with β extension agreed tolerably well with calculation based on the assumption of rotation of glide planes within the grains. Extension of cadmium at −180°C. and −78°C. led to abnormal results, traced to extensive twinning that takes place at low temperature. Thus everything indicates that resistivity measures glide plane rotation in non-cubic crystals and shows that β flow is connected with slip, closely coupled with rotation, within the grains.

In connection with this work Gibbs and Ramlal measured directly the mean change of angle by X-ray transmission methods. They found no change during an immediate extension of four per cent and very little during extension after thirty per cent, when β flow had been almost exhausted, but considerable change of angle in the intermediate period of β flow. In general, their results supported the resistivity work.

As regards the κ flow, the close connection between this and grain boundary is well brought out in some fundamental work by J. McKeown, who carried out, at room temperature and constant stress, long-time (300 days) creep tests with specimens of lead of mean grain areas from 0·84 to 0·01 mm². At one fixed stress he found, firstly, that while the creep with the coarsest-grain material showed well-marked β flow as well as κ flow, the *relative* importance of β flow diminished and that of κ flow increased as grain size diminished, until with the finest grained metal the flow, practically linear with time from the start, was much more rapid than the final flow with any of the other specimens. Secondly, the κ creep rate was inversely proportional to the mean grain diameter, that is, proportional to the area of grain boundary per unit volume. Other work at the British Non-Ferrous Metals Research Association has shown, though not in the same detail, that grain size has a similar effect on the creep of lead alloys, copper and aluminium bronze. These grain size experiments, then, support the view that the κ flow is mainly, at any rate, due to processes taking place at the grain boundaries.

Before considering work carried out in the absence of precise creep measurements or analysis of the creep curve, it may be well to point out that, generally speaking but not invariably, high temperatures and slow rates of strain can be taken together as favouring easy attainment of a given deformation, mainly by κ flow: low temperature and rapid rate of strain imply difficult extension, mainly by β flow*. I propose to call the former 'yielding conditions', and the latter 'stubborn conditions'.

* e.g. with lead, for a given β, not too high, κ is about thirty times as great at 162°C. as at 17°C.: at high values of β, κ is still much greater at the higher temperature, although the difference is not so marked.

Hanson and Wheeler carried out important work on the metallographic features of creep, in which they examined microscopically at the metal surface the grain behaviour associated with different kinds of creep behaviour, a procedure that is open to the general criticism that, as will be pointed out, the behaviour of surface grains is not necessarily characteristic of the bulk of the metal. With polycrystalline aluminium at room temperatures they observed coarse slip bands to appear in the grains as creep proceeded: rapid strain at all temperatures up to about 400°C.—'stubborn' conditions—similarly produced plentiful slip bands. Slow straining at 250° or higher temperature—'yielding' conditions—gave rise, however, to no visible slip bands until considerable extension had been produced, and then the bands were faint and fine, of a different character from the familiar bands. At 400°C. no slip was detected. The crystal boundaries were, however, very much intensified in these high temperature experiments, especially after considerable extension had taken place and ultimately actual cracks appeared between the crystals, which led to fracture. It may be noted that Hanson and Wheeler clearly say that when so-called 'slipless' deformation takes place, this only means that the normal well-marked bands are absent, but they bring evidence for the creation, under these conditions, of a large number of planes with very small slip on individual planes, which makes them difficult to observe: slip is 'so well distributed as to be invisible under the microscope'. They gave, as the conditions favourable for this type of slip, slow rate of strain, high temperature and small grain size—the conditions that favour κ flow. In view of recent work this clear enunciation in 1931 deserves to be stressed. Hanson and Wheeler's results, then, although they did not so express them, connect clearly β flow with coarse slip bands within the grain and little boundary flow, κ flow with very fine bands within the crystal grains and marked boundary development and flow.

The flow at rapidly increasing rate, known as tertiary creep, may have a variety of causes, most of which lead to very similar forms of strain-against-time curve. In many of the earlier experiments the increasing rate was undoubtedly due to the increase of stress which takes place in tension experiments carried out under constant load and was completely eliminated with a constant stress device, as I showed with lead, for instance. Tertiary creep due to increase of stress is trivial, can be avoided by working under constant stress and presents no problem. But there is real tertiary creep, due to structural changes, as convincingly shown by Sully and his collaborators in experiments which demonstrated tertiary creep in compression.

Recrystallization is a fruitful cause of tertiary creep, and I here use the term to cover crystal recovery (also called crystal relaxation), growth of new crystals from nuclei, and boundary movement or grain growth. Phase change is also a possible cause of accelerated creep: for example, this behaviour in an austenitic chromium-nickel steel has been attributed to change from a single phase austenitic structure to a duplex austenite plus ferrite structure. In general it is clear that any structural change in which atoms are changing their allegiance from one system of ordering to another must lead to increased mobility, since

at the moment of transfer they are not built into a stable (or metastable) structure.

The first demonstration of recrystallization as productive of accelerated creep was by Greenwood and Worner, who established that recrystallization could take place under stress with lead at room temperature. They worked at constant load and reached elongations of some per cent, so that the last stages of acceleration are probably due to increased stress. In general, recrystallization under stress appears to take place fairly suddenly. Sinclair and I followed the behaviour of very pure leads, which recrystallize easily under stress but are quite stable when unstressed, at temperatures from 0° to 100°C. The increase of rate, which was clearly shown by microphotographs to be connected with recrystallization, was followed by a decrease associated with a completion of the recrystallization process and the curves at lower stresses gave the appearance as if a new lead had been formed, which then gave a normal type of creep curve. We found that, within the temperature range in question, recrystallization took place at a characteristic strain, which can be easily explained. There is another type of accelerated creep, illustrated by Tapsell and Remfry's results on a high-nickel high-chromium steel at 800°C., which is characterized by a rapidly increasing acceleration.

Tertiary creep can also be produced by transverse intercrystalline cracks which appear, as shown by Hanson and Wheeler, and by subsequent workers, at a late stage of the deformation. It is clear that these are bound to promote creep in an ever-accelerating fashion, by throwing additional stress on the sound structure in a way that produces still more cracks. Tertiary creep due to this cause will lead to speedy intercrystalline fracture.

Enough has been said, perhaps, to show that one cannot speak of *the* cause of tertiary creep. There are a number of possible causes, of which two or more may be operating at the same time. In the old tests under constant load the trivial increase-of-stress effect is always imposed on any other causes that may be effective. It should be important for the practical man, who is trying to eliminate tertiary creep, to ascertain what is the effective cause of it in his tests and not to declare, because one has been traced, that no other can be in question. Increase of stress, recrystallization and intercrystalline cracks are not mutually exclusive.

This brief survey of certain features of creep that may, I think, be taken as established suffices to show reasons for some of the intricacies that trouble the worker in this now popular subject. Consider small impurities, traces of foreign metal. They may segregate preferentially to the disordered boundary region: whether they do or not is partly a question of such factors as atomic size and atomic field. If they do, they will have a disproportionate influence on the κ flow. Boundary segregation is well illustrated by polycrystalline cadmium containing a trace of bismuth, which gives intercrystalline fracture under very small stress at 100°C. The complications which they may introduce if built into crystal structures, that is, if they remain within the grain, are well known to vary according as they do or do not fit well into the crystal pattern. But these matters I have

expressly avoided, more especially since the less that is known about a subject, the longer it takes to discuss. The various possible modes of deformation of the single crystals of the grains, not only by pure glide (which in the case of cubic crystals involves a variety of glide planes and glide directions) but also by other mechanisms to which I shall have to refer, do not simplify matters, more especially as they react on the boundaries.

Confining ourselves to relatively pure metals, then, which removes many of the problems in which many of you are mostly interested, we are confronted with the difficulty that, throughout a specimen, the crystal grains do not, even when averaged over limited regions, deform in the same way. Grains at or near a free surface, in tensile or compressive tests, behave, as a class, quite differently from grains within the interior. As far as I know, this was first emphasized by Andrade and Kennedy in a brief account of the creep of wires of different diameter under tension. The metal was lead containing a trace of tellurium, which at atmospheric temperature shows transient creep only. Wires in which the diameter is comparable with the grain size creep, under a given stress, more readily than those of larger diameter and the effect can be expressed by the formula

$$S = S_0 r^2 (r - na)^2,$$

where S is the stress required to give a particular value of β, S_0 is a constant giving the stress for that value of β with thick wires, r the radius of the wire, a the average grain diameter and n a constant which turns out to be 0·3. Thus the effect can be formally accounted for by supposing that an outside layer of the wire, a third of a grain thick, has no creep resistance at all, meaning, of course, that the outer grains are comparatively weak. This diminished creep resistance of outside grains can be easily understood, since at the free surface there is a lack of the geometrical constraint which hinders deformation of inner grains. Etching away the outside of a wire, rapidly extended by ten per cent, by five grain diameters showed, on X-ray examination, a complete change of crystal structure as compared with the surface. This is a warning that the examination of the surface during or after tensile creep may tell very little about grain deformation in the main body of the metal. It is also a reminder that a free surface may greatly affect creep.

The difference between the behaviour of surface grains and inner grains has been brought out in work by Rachinger with pure aluminium. He found that, with a specimen slowly strained at 300°C. the elongation of the individual grains at the surface was twice as large as that at about one grain deep and five times as large as that two grains deep.

Quite apart from surface effects, individual grains are not only oriented differently but are subjected to quite different stress distributions. One would sometimes imagine in reading papers on the subject that an individual grain, of uniform cross section a in the plane normal to the axis of the specimen, cross section A, was subject to a uniform tension a/A times the applied tension. Actually each surrounding grain is tending to slip on differently oriented planes in different directions, and there may be strong compression or shearing forces on the grains in question. Further, the plastic deformation of an axially extended

cylinder is, owing to the contraction, not simple but complicated. The work of Chang and Grant has shown particularly clearly the difference of behaviour from grain to grain. More than this, Boas and Hargreaves have shown that, as might be expected from the irregular stress field, the behaviour within a crystal grain does not show the regularity exhibited by a single crystal as usually stressed for measurement. The hardness and the elongation vary as the boundary is approached and the coarse slip bands often stop short of the boundary.

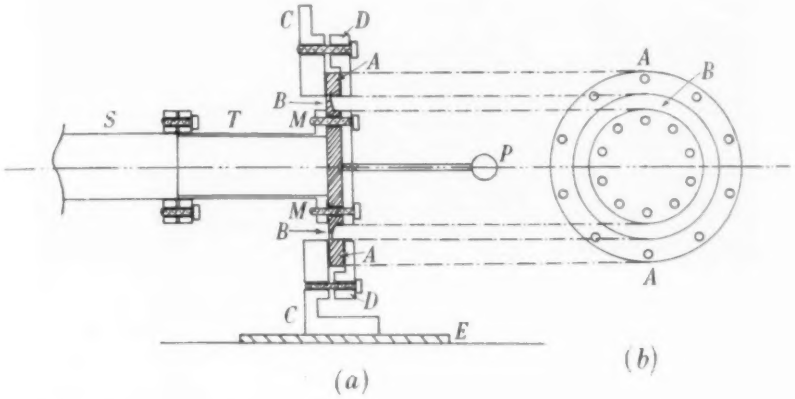

FIGURE 1. *Experimental arrangement for creep under simple shear*

Many of these points are brought out in work by Andrade and Joliffe on the creep behaviour of metals under simple shear. The set-up is made clear in Figure 1: the metal is in the form of a disc, in which an annulus is cut, the outside part of the disc being gripped and the inside subjected to constant torque, so that the thin part of the metal is subjected to constant shear. In these circumstances it was found that the metals concerned—pure lead, commercial lead and cadmium—gave a pure $t^{\frac{1}{3}}$ law for shear strain against time, without any permanent creep, even at temperatures where, in ordinary tension experiments, permanent creep was very marked. With pure lead recrystallization under stress, accompanied by rapidly accelerated creep, was observed and, as in our tensile tests, it was found that this took place at a characteristic strain. The absence of permanent creep I attribute to the fact that under these conditions the grains at the free surface have very limited freedom as compared to the tensile case, where the area of the free surface, here constant, is changing rapidly. The conditions in the shear experiments cannot be very different from those in a sheet of thickness infinite in the axial direction. The geometry is unfavourable for grain boundary movement. The slip bands, and their ending short of the boundary, is shown in Figure 2.

This shear disposition allows us to study conditions under reversal of directions of motion, which has led for the first time to a marked contrast in creep

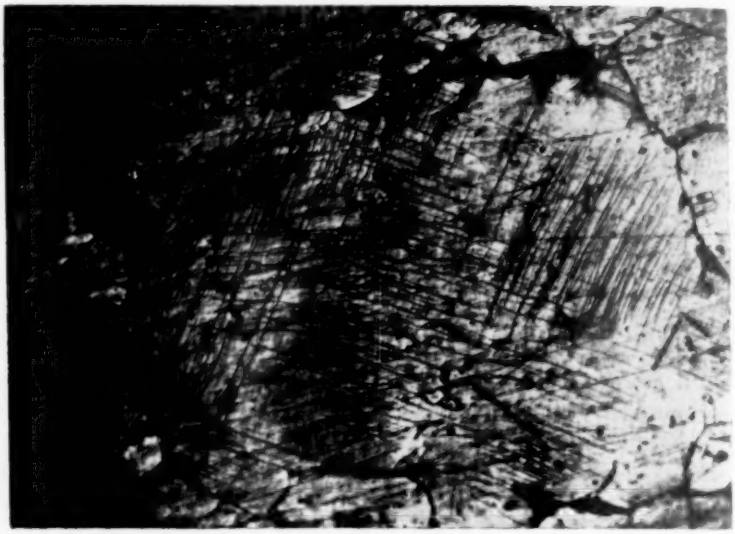

FIGURE 2. *Glide within grains in direct deformation of commercial lead under simple shear. Shear strain 5 per cent*

behaviour between cubic and hexagonal metals, as shown in Figure 3. In the case of both commercial lead and cadmium the creep under first direction of shear follows closely the $t^{\frac{1}{3}}$ law. With cadmium it follows the same law under reversed shear and under a second reversal: with lead it gives shear strain linear with time in both cases. This can be explained by considering that glide in lead single crystals leads to great hardening (typical of cubic metals), while that with cadmium leads to little hardening*, and further that with lead there are four possible glide planes, each with three possible glide directions, as contrasted with one glide plane, with three possible glide directions, for cadmium.

During the forward shear strain of some twenty per cent the lead grains harden considerably and the multiple glide leads to much intergranular locking, which hinders boundary movement. Under reversed shear there is practically no movement within the hardened grains, but intergranular movement in the reverse direction is possible, owing to boundary fragmentation which, however, did not allow forward κ flow owing to interlocking. The boundary movements create further disruption, so that linear time flow is easier on second reversal. It is also significant that with the pure lead there is recrystallization under forward shear but not in reversed shear. With cadmium in forward shear there is little hardening in the grains owing to glide, but considerable geometrical hardening owing to limited glide planes. Hence the process in reverse is very

* At atmospheric temperature, critical shear stress for lead is 14 g/mm^2: stress at glide strain of 0.3, 450 g/mm^2. The corresponding figures for cadmium are 50 and 80 g/mm^2.

similar to the forward process, of which it is merely a repetition. I should like to see further metals tested in this way over a range of temperatures.

Recent work has revealed that the changes within the grain during creep may be complicated by effects that we may class generally as sub-crystal formation.

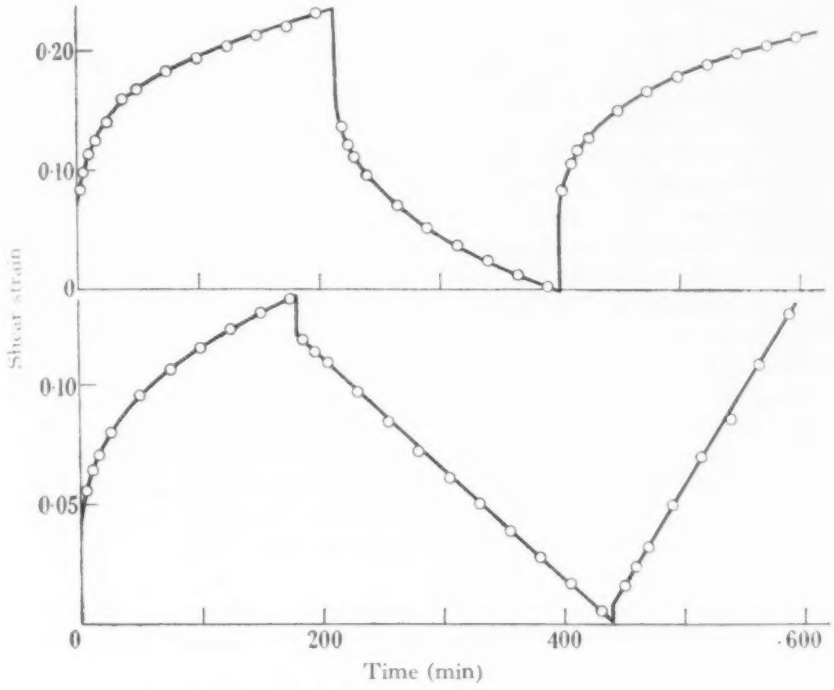

FIGURE 3. *Creep under simple shear with two reversals: (above) cadmium at 140 kg/cm²; (below) lead at 53 kg/cm²*

In this connection I may, perhaps, be allowed to recall that Chow and I, working with single crystals, described in 1940 work demonstrating that the disordering of crystals by local rotations, the process later elucidated by Cahn and christened 'polygonization', was closely connected with hardening and that at higher temperatures the disorder produced by a given strain diminished, while Henderson and I, extending the results, emphasized that when glide was concentrated in comparatively few regions (widely spaced coarse slip bands) the disorder corresponding to a given strain was greater. These facts are involved in some of the most recent work.

Fine slip is a process which has been revealed by the electron microscope. With this instrument Heidenreich and Shockley have demonstrated, with aluminium, that the familiar coarse slip bands are a cluster of fine bands, the elementary glide planes being about 2×10^{-6} cm. apart and the glide taking

place on each being about ten times this amount. Using similar methods Wilsdorf and Kühlmann-Wilsdorf have found a very fine surface structure—with aluminium, of course—parallel to the coarse slip bands, with about the spacing just given, and McLean, using the phase contrast microscope, has likewise demonstrated the existence of very fine slip bands between the coarse slip bands. An important result is that fine slip is not associated with disordering and hardening. It is, in general, responsible for only a small part of the strain and possibly it may take place without rotation of the planes, in which case it would account for certain small discrepancies in the results of Andrade and Chalmers.

Fine slip is, then, a process that can take place in the grain. Further McLean, in agreement with Honeycombe, Crussard and others, has shown that part of the crystal lattice may rotate about a line in the glide plane normal to the direction of glide, so that the planes are, locally, continuously bent. Such bands of local rotation are known as deformation bands. Or the planes may become distorted into small portions, each of regular crystal structure, of which the planes are tilted successively in one direction and the other, the polygonization already mentioned. Figure 4 shows diagrammatically how McLean considers the structure to arise, with aluminium. The crystal planes in the grain must, then, not be thought of as a neat pack of uniform cards uniformly sheared. McLean, from careful surface measurements of coarse slip bands and grain boundary displacement, found them insufficient to explain the observed over-all extension, and accounted for the difference by slip on the fine slip bands. This, of course, on the usual assumption that the surface is typical of the whole.

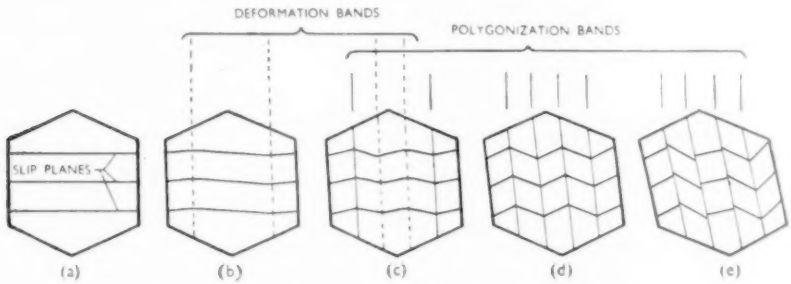

FIGURE 4. *McLean's picture of lattice deformation during creep: (a) initial condition; (b) first stage creep; (c) first-second stage; (d) second-stage creep; (e) late second-stage creep*

A considerable body of work directed to correlate changes within the crystal grain with creep processes has been carried out by W. A. Wood and his colleagues in Australia. Here, again, we are met by the difficulty that all observations were on the metal surface, since, although X-rays played an important part, the method used was back-reflection of the easily absorbed cobalt Kα rays. I think that it is, perhaps, a pity that the method of transmitted rays is seldom used

to-day. A further difficulty is that the extension was carried out at constant rate, which means that, in general, the stress was being continuously raised during the process. In describing very briefly the results, the distinction which I have made between 'yielding' and 'stubborn' conditions will be used, which is compatible with the method.

With the disposition adopted, the X-ray pictures of the surfaces of the undeformed metal showed well-defined spots, which means that within the area of the beam there were a few, well-defined grains. Strains under stubborn conditions produced, in place of the spots, continuous arcs, indicating a continuous range of angles of crystal deformation within the grains. On the other hand, under yielding conditions the individual spots were split into smaller discrete spots, indicating break-up of the grains into perfect-crystal portions of slightly different orientation. Correspondingly, microscopic examination showed coarse slip bands in the first case, and revealed small crystal units within the grains in the second, to which Wood has given the name 'cells'. There may be ten or more such cells in a grain. With intermediate conditions the slip bands were coarser and sparser, while at the highest temperature no slip lines at all were seen, but very marked crystal boundaries. High temperatures and slow rates—very yielding conditions—favour large cells, in the limit cells equal to the grains, that is, no division into cells. Cell formation is now a recognized possible form of grain deformation and has been detected in zinc and magnesium as well as aluminium, but on the other hand it has been sought for in vain in lead and certain other metals.

A brief account has been given of the processes within the grains to indicate the complexity of the problem when it comes to details of disturbances, a complexity which is stressed when the diversity of opinion between workers in the field, for example McLean and Wood, as to the interpretation of the observations, is considered. Coarse slip, which is merely fine slip concentrated into narrow domains; fine slip; deformation bends; polygonization; kink bands all come into the discussion. As, however, all this is surface behaviour it is not certain that detailed discussion will necessarily clear up the creep problem. What seems established and agreed is that under stubborn conditions, which I connect with β flow, there is considerable formation of coarse slip lines within the grains, which is associated with local angular deviation of lattice planes and with crystal hardening; that under β flow conditions there may be other types of grain deformation: that under extensive yielding conditions—prevalence of κ flow—there is absence of coarse slip lines but abundant fine slip, not associated with hardening: that under flow that are marked boundary movements. All this agrees with the general picture that has been given. When it comes to reconsidering details of the mechanism of metallic creep, evidence from surface measurements on one metal, aluminium, is not sufficient. Twinning, for instance, which I have had no time to treat, is of frequent occurrence in stubborn conditions with some metals, for example cadmium, especially under simple shear.

The question of boundary movement is, like that of grain deformation, a very

complicated one: this is bound to be so, for the two are very closely connected. One crystal grain clearly cannot move with respect to other grains without a certain amount of deformation of the grain itself and of deformation of the boundary: only certain surfaces of great simplicity and regularity can slide past one another if rigid bodies are in question. Chang and Grant have established that a migration of boundaries can take place during creep. It has also been shown that, as might be expected, boundary motion is not uniform in time or space. Fine slip gives an easy method by which the grain can modify its shape, in the course of boundary movement, without marked hardening. The study of surface grains, as normally carried out, does not seem well adapted to tell us of the behaviour of boundaries within the metal. It is suggested here that the study of surface grains in experiments of the simple shear type might be instructive, in forward and backward shear, which enables the same grains to be observed both under conditions that do not give, and that give, κ flow.

We now come to consider briefly the theoretical aspect of creep. Much work, of great ingenuity and elaboration, has been done by way of applying the theory of dislocations, originated at Frank-Read sources, to the transient flow. Mott, in particular, has introduced the conception of climb from a blocked dislocation ('jogs') as a process of relief which enables further deformation to take place and leads to deformation bands and the so-called polygonization. On the assumption of an exhaustion of dislocations he deduced, with Nabarro, a creep logarithmic with time, which is what Wyatt has found at low temperatures and very slow rate of creep. In such cases there is no recovery and an increment of stress during creep produces an effect independent of previous creep—in other words the metal, exceptionally, obeys an equation of state. With special assumptions concerning the escape of dislocations from piled-up groups, Mott deduces the $t^{\frac{1}{3}}$ law.

It seems to me that the machinery postulated in such attacks on the problem is, possibly, over elaborate, and the assumptions made a little too particular, when it is a question of explaining so general a law as the $t^{\frac{1}{3}}$ law which, over a wide range of stress and temperature, has been verified by independent observers with cubic metals, hexagonal metals, rhombohedral metals, tetragonal metals, pure and impure, in some cases over periods of weeks and also for a variety of non-metallic substances. At the time when I ceased to have facilities and circumstances favourable for research, I was working on a theory on a general basis, which I still hope to bring to a stage of completeness. The first assumption is that the creeping substance contains a number of flaws—dislocations if you will—of various severity: more precisely, each flaw is characterized by a stress τ which causes it to propagate a limited strain disturbance s, and τ is distributed statistically. One piece of evidence for flaws of varying severity is the way in which the spacing of coarse slip bands in metal single crystals increases with temperature. At high temperatures, where glide continues readily, only the few flaws characterized by low values of τ are started, and then run so easily that the higher stresses needed for the more stubborn flaws cannot be applied. The strain produced by yield of a given flaw is assumed to be a very simple

function of τ, more stubborn flaws producing bigger strains when they yield than do the less stubborn flaws.

When a stress is imposed, then, all flaws up to the local value of stress caused by the external stress yield, and produce an immediate extension. The values of f_τ given in my early papers for lead, copper and iron at different temperatures lie on one curve, when the scales are suitably modified and this curve agrees well with that deduced from a Gaussian distribution of τ and exhaustion.

Transient creep then takes place for two causes, both activated by local temperature agitation. Firstly, flaws characterized by values of τ larger than the limiting value for immediate extension are activated by local increases of stress due to temperature fluctuations, as assumed by R. Becker and Orowan in their treatment of plastic flow. The yield of each flaw makes, by the general law assumed, a larger contribution to the strain than any of those flaws which yielded immediately. Secondly, flaws which yield at once, run a given distance, and are then blocked, giving the immediate extension, are enabled by a favourable temperature fluctuation to make a further contribution to the strain. The details of the machinery may well vary from case to case: for instance, in cases where the coarse slip bands do not reach to the grain boundary—see, for example, Boas and Hargreaves, and Andrade and Joliffe—they could probably advance further under continued constant bulk stress. The limited fine slip lines of Wilsdorf and Kühlmann-Wilsdorf can probably extend too by this mechanism.

The preliminary work indicates that, with a suitable and reasonable choice of constants, the complicated expression reached can give a close approach to $t^{\frac{1}{3}}$, the small departures being, in the numerical case worked out, in the very early stage of creep. A theory that will, according to the choice of constants, give a little range of form would seem to be well adapted to express the small variations from $t^{\frac{1}{3}}$ flow—for instance, values of the exponent slightly different from $\frac{1}{3}$—which have been recorded.

The theory leads to a variation of β with stress similar to that which I and my collaborators have found, expressed by a sigmoid curve on which, after rapidly increasing with stress, the rate of increase of β decreases markedly. For, speaking generally, if a large proportion of the flaws have been activated in the immediate extension, the first effect of temperature agitation, recorded above, will not be so pronounced, but the second, of course, persists.

As regards the κ or persistent creep, the law that log κ is a linear function of the stress, to which I drew attention many years ago, has been explained by Kauzmann, Dushman and his collaborators, and others on the basis of the viscosity relation $\eta_i = Ae^{b/kT}$. The constant b, which I originally called just the viscosity energy, has been renamed, after Eyring, the activation energy and, from this, exponential expressions, either in terms of hyperbolic sines or the simple exponential, have been derived for the variation of κ with stress and temperature. It is implicitly assumed that what we are dealing with is, in effect, a sheet of homogeneous—speaking in macroscopic terms—material, of constant thickness, under shear. Such may reasonably be taken to be the conditions in the experiments of Zener and Ke at low stresses, which lie outside the range normally

considered in creep: in these experiments, however, the rate of slip is proportional to shear. On the other hand, the disordered intergranular matter which is responsible for κ flow in normal creep experiments cannot reasonably be assumed to be of thickness constant for all values of stress: on any of the prevailing views the thickness must increase with stress, either by the accumulation of the effect of dislocations which have run out of the grains, by boundary migration or any equivalent mechanism. In any case, microscopic observation is sensitive enough to show a broadening of the boundaries in extreme cases. It is partly the β flow which leads to boundary disturbance, and here an interconnection between β and κ flow, on which Orowan has laid stress, comes in. As β flow, over a wide range, increases markedly with stress; as boundary disturbance due to other causes must likewise increase with stress; and as the rate of flow in a stressed disordered sheet is proportional to the thickness, a rapid rate of increase of κ with stress is to be anticipated, independent of changes in activation energy, but of this no account has been taken.

I have endeavoured to set before you certain general considerations concerning the nature of the creep process, particularly with a view to separating what has been experimentally established from the hypotheses as to what is occurring to dislocations that are still insufficiently identified. In the course of preparing this lecture, I have often regretted that modern university and government policy—as to the wisdom of which I have no doubt—denies to a man who has reached the age of 65 both laboratory facilities and such conditions as are in any way favourable for concentration on a scientific problem. There are various experimental investigations which would, I think, considerably help us towards understanding the creep process in simple metals—and I repeat that I regard this as the first step towards a more general understanding. I should like to have carried out much more work under conditions of simple shear, where I think the behaviour of surface grains more characteristic of internal behaviour than in the customary tests. Tests at long times of flow and for a variety of typical metals, including tin—and, of course, aluminium—at a range of values of θ could be instructive. I should like to have followed the effect of repeated reversals with cubic and hexagonal metals, more particularly to find out the final effect on creep behaviour. The behaviour of given selected groups of grains in a cubic metal under forward shear, when the creep is pure β and under reversed shear, when it is pure κ, might repay study. I should like to have tried the effect of forming wide disordered boundary regions, by κ creep under large stresses, on subsequent κ creep under a range of small stresses. I should like to have investigated the behaviour of the coarse slip bands in β creep proceeding under simple shear with typical metals—and a dozen other points.

Under tensile testing, the influence of a wide variety of surface conditions is an attractive subject. But these things are not for me. I can only thank you for this stimulating opportunity of talking to you about these problems that interest me so much, and for the encouragement afforded by your patient hearing.

III. FATIGUE

by

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Monday, 28th March, 1955

It is not customary to introduce a Cantor Lecturer. I must therefore do this myself. I will be brief, accurate and non-laudatory. A Chairman can seldom be these three things and sometimes he can be none of them. I am a Metallurgist, which is almost equivalent to saying I am a member of the depressed classes. For considerably more than a generation, I have been using my metallurgical knowledge in the ever-expanding aeronautical field. I am going to speak now of the fatigue of metals—to try to put before you all the important facts in relation to fatigue in the course of a lecture of an hour's duration is a task comparable with that undertaken by the late Mrs. Partington, who attempted to keep out the Atlantic Ocean with a mop—it is on record that she failed, and I am not expecting to succeed.

Since the fatigue of metals has been studied, about 1,200 worth-while papers have been published with regard to it. At the best, I can only hope to give something of a sketch—under the circumstances a picture, even in the current slap-dash style so popular for presentation portraits, is quite beyond my powers. Unlike Gaul, my lecture is divided not into three, but into five parts:

(1) The history of the discovery of fatigue,
(2) The mechanism of fatigue,
(3) The relationship of stress to endurance,
(4) Some disturbing facts, and
(5) A summary.

PART I. THE HISTORY OF THE DISCOVERY OF FATIGUE

During at least the last five thousand years, Man, and here man embraces woman which is not infrequently the case, has come to know an ever-increasing amount about metals. This knowledge has been almost wholly derived, not from theory, but from experiment often unwittingly undertaken.

Fatigue of metals, which now so rightly fills with dread the minds of designers of mechanisms or structures on which human safety depends, was admittedly a late discovery; but in spite of what the popular press might lead one to suppose, it is by no means a contemporary one. About a century and a half ago Albert, a German engineer of whom we know little, noted that wrought iron chains used in mine hoists, after a period of satisfactory service, broke with a brittle fracture, like that of china, under a load well below their proof one and not in excess of their normal one.

In France, somewhere about the same time, the breaking of wrought iron tyres of the wheels of stage-coaches was observed, and later investigated by two

engineers, Marcoux and Arnoux. They noted the brittle fractures, characteristic as they say of iron of bad quality, but they were satisfied that the iron in question was not bad *ab initio*—it had become bad or brittle in service—the terms were to them synonymous. They also noted that use had not produced a general, so much as a local loss in the ductility of the metal of which the tyres had been made.

Finally, in England only a little over a century ago McConnell, in papers given before the then very young Institution of Mechanical Engineers, reported that railway axles failed, under normal loads, after long periods of satisfactory service. Like his German and French fore-runners, he observed that, though the actual fracture was like that of brittle material, the metal away from the fracture possessed ductility comparable with that which it originally had.

So much for an all too condensed history of the discovery of fatigue.

PART II. THE MECHANISM OF FATIGUE

Now what was it Albert, Marcoux, Arnoux and McConnell saw, even 'as through a glass darkly'? In a way, this is shown in Figure 1. Figure 1 shows the side view of a bit of the floating axle from a small motor car, after the axle had ceased to float. The coin on the right is a halfpenny of mine—it is there to give an idea of size and not as a vulgar and ostentatious display of personal wealth. It would not be slanderous to say that the shaft had broken off like a carrot.

FIGURE 1. *Left, a portion of the floating axle of a small motor car. Note the axle has broken with a brittle fracture while, right, the tensile test piece, cut from as close as possible to the fracture, shows marked ductility*

The second view on the figure is of a testpiece made from the axle, very near to the fracture. This testpiece has been broken in tension. You will notice that the material shows very marked ductility when broken in this way—the metal has stretched before breaking. Its mechanical properties, neglecting decimals—'those damned dots', as Sir Winston Churchill's father used to say, were:

Proof Stress	65 tons/in.^{2}
Ultimate Tensile Stress	71 tons/in.^{2}
Elongation	20%

In Figure 2 there is another view of a similar and largely brittle fracture: this is the broken steering drop arm of a popular ten horse-power car—of course, after the break, it was not too popular with its owner. It will be noticed that there appears to be a dividing line across the fractured area. I will refer to this later.

Figure 3 gives yet another example of failure, not, in this case, provided by the unwitting generosity of the Motor Industry: this shows the fracture of a seven-inch diameter mill shaft, which had run satisfactorily for years and broke when the mill was operating under perfectly ordinary conditions. Note the non-uniform character of the fracture—the centre area is different from the remainder.

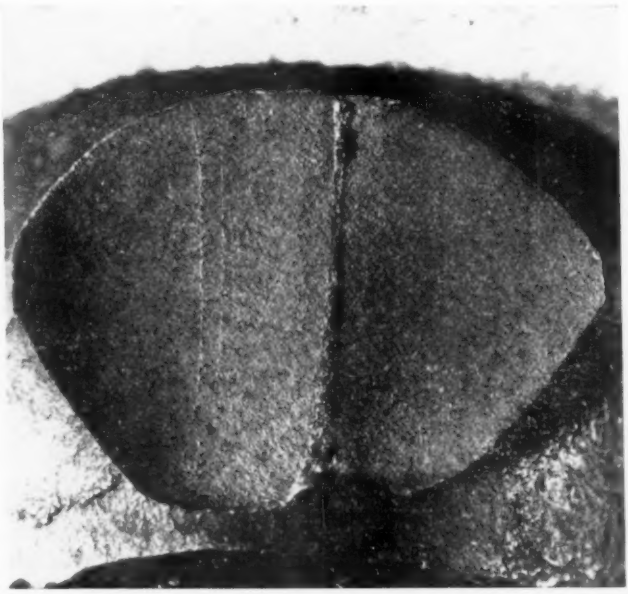


FIGURE 2. *Motor car steering drop arm. Note the two types of fracture*

Figures 2 and 3 each illustrate two types of fracture—brittle and ductile. Separation in the smoother area took place slowly, but as inevitably as the march of time. The relatively rough zones are characterized by evidence of plastic flow. Fracture occurred here with dramatic suddenness. The hitherto unbroken metal, unable at long last to sustain its oft-repeated load, gave way. To summarize—in a period possibly starting prior to the French Revolution and going on into our industrial one, a few observant engineers had established that parts made of ductile metal sometimes broke, after repeated use, when subjected to loads which they had long withstood with apparent immutability.

Poncelet, the French engineer, pondering on this phenomenon—he probably was something of an expert on pondering, as he was in a Russian prison from 1812–1814, where he doubtless had ample opportunity for what Mr. Nehru, with his considerable experience of imprisonment, calls unheated reasoning—christened, or better, for so evil a thing should surely not be christened, gave it the name 'Fatigue'. McConnell, rashly daring, produced a theory to explain it. This has long been destroyed as have many others which have succeeded it—in this field, the theorist has been akin to: 'The Priest who slew the slayer and shall himself be slain', a fine phrase from Macaulay, but perhaps too good to apply to a theorist?

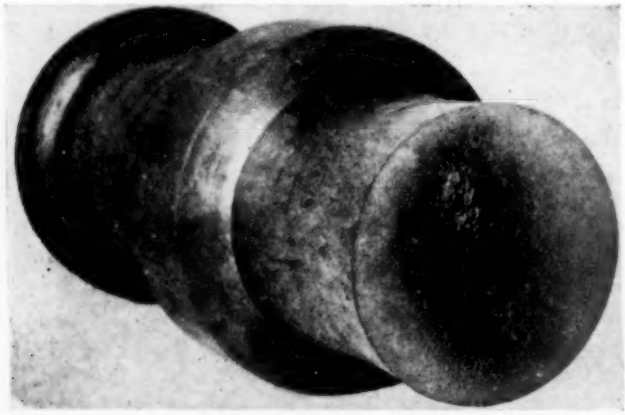


FIGURE 3. *Millshaft. Note the two types of fracture*

Observations as to the conditions giving rise to the largely brittle failure of ductile metals were made in many countries but nowhere more thoroughly than in Germany. Here, in the ten years prior to the Franco-German War, Wöhler produced a series of outstanding papers from one of which it is now appropriate to make a quotation in translation: 'Wrought iron and steel will rupture at a stress not only less than the ultimate static strength of the material, but even less than the elastic limit, if the stress is repeated a sufficient number of times.' This was a great contribution to knowledge, for it had hitherto been believed that if the stress imposed upon a piece of iron or steel was below its elastic limit, it would withstand this indefinitely—Wöhler had shown that this was not so with the ferrous materials he tested. It was only true if the stress was static. He had established, beyond a peradventure, a property of repeated ones—and be it marked, there is no qualification as to the type of stress, the emphasis is upon its repetition—it might be a tension, compression or shear one or a combination of these.

While Wöhler had shown what repeated stresses did to metals, neither he nor his contemporaries had any idea as to how they did it. Metallography was

in the pangs of childbirth: X-rays were not yet conceived and were to remain unborn for half a century. These two interesting events having long taken place, we, as heirs of the past, can now examine with ever-increasing understanding the surface of a piece of metal, both by the optical and the electron microscopes, while it is being subjected to repeated stresses—we can also look beneath its surface by the use of X-rays.

As a result of this elaborate emulation of 'Peeping Tom', we can not only see the metals naked and unashamed, but also many of their innermost secrets. We now know all metals are composed of crystals—they have much in common with a lump of sugar—and all crystals are composed of atoms arranged according to one of 14 different patterns or lattices. Figures 4 and 5 do something to illustrate these facts.

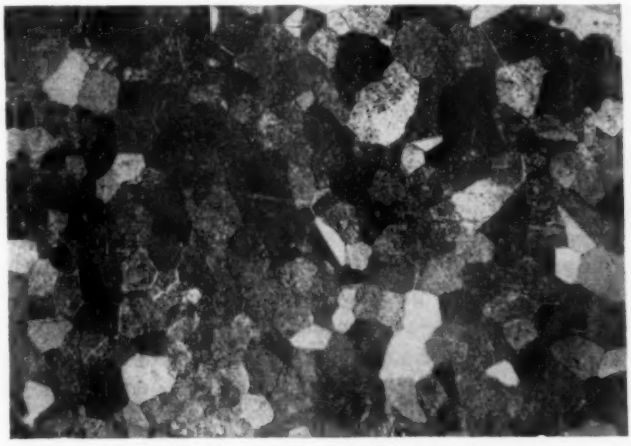


FIGURE 4. *Photomicrograph of pure cast gold. Magnification approximately three diameters*

Figure 4 is a photomicrograph of the polished, etched surface of a piece of annealed pure gold— it contains less than one part in 100,000 which is not gold. Now what does one see? If you are not a metallurgist then you see something like a patchwork quilt made of patches which have been produced by tearing rather than by cutting; if you are a metallurgist you see allotriomorphic crystals separated by intercrystalline boundaries, whose thickness is estimated to be a few millionths of an inch.

Allotriomorphic is just one of those words by which the expert attempts to mystify the non-specialist. Doctors have something of this technique, what they describe as a contusion of the aural epidermis is really a thick ear. To return to the allotriomorphic crystals—admittedly the two-dimensional view in the figure gives no impression of crystals, but as these have all grown simultaneously they have poked into each other, thus eliminating the external characteristic plane faces of crystals which have, as it were, been able to grow up

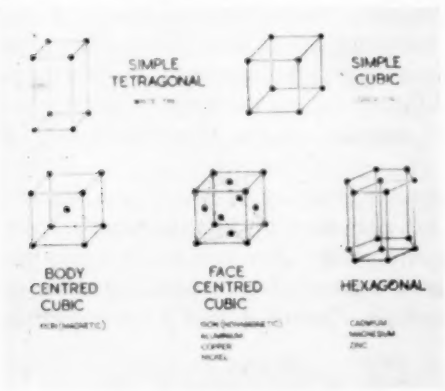


FIGURE 5. *Distribution of atoms in the lattice of some metallic crystals*

like a stately home in the open country rather than in the confusion of a crowded caravan site. Of what are these crystals made? The answer is of atoms, arranged according to one of the 14 patterns or lattices already mentioned. Figure 5 shows the lattices of some of the metals. The spots represent the atoms. The lines merely indicate distance and have no existence in reality. In the case of aluminium and of gold, the shortest distance between atoms is about one hundred millionth of an inch. In the average gold crystal shown in Figure 4, there are about 6×10^{19} atoms.

While the metal is in the molten state, these atoms have considerable freedom of movement and are akin to troops milling about the parade ground. When the metal has solidified, their positions are analogous to those of the men immediately after the order 'fall in' has been given. Some atoms have not had time to get to their proper places—there are imperfections in their ranks—in the crystal, the metallurgist is less effective than the sergeant-major in dealing with irregularities similar to those which occur on the parade ground. A vast amount of imaginative work has been devoted to evaluating the significance of these imperfections within a crystal. Due to lack of time I must dismiss this—not indeed with contempt, but with a couple of sentences. These speculations have given rise to many ingenious ideas, some of which tax and perhaps overtax the credulity even of those who find no difficulty whatever in swallowing the story of Jonah and the Whale.

So much for the structure of a piece of metal. We can now ask a question: do repeated stresses change this structure? The answer is definitely yes. This is illustrated in Figure 6, which is due to Gough (himself a Cantor lecturer) and to Hanson. The magnification of both parts is 1,400 diameters. The upper portion, which may look like a first-class view of next to nothing, is a photomicrograph of the surface of a polished specimen of a singularly pure piece of ARMCO iron. One sees merely a couple of faintly indicated intercrystalline boundaries. In the bottom part of the figure, the same piece of metal is shown under the same degree of magnification, after 40,000,000 stress reversals of $\pm 12{\cdot}2$ tons in.$^{-2}$. The appearance of the surface as seen under the microscope has changed—a countless number of what are called slip planes or slip bands has developed. The lines are due to minute differences in local levels on the original surface. The analogy is a poor one, but the surface of a crystal can be likened to a pack of cards seen edge on. In the upper part of the figure the edges of all the cards lie in the same plane, in the lower part the cards have moved relative to each other and there are, as a consequence, a series of minute steps.

To revert to the slip band—it has come into being, owing to the repeated stresses having produced permanent movement on the atomic planes of a crystal.

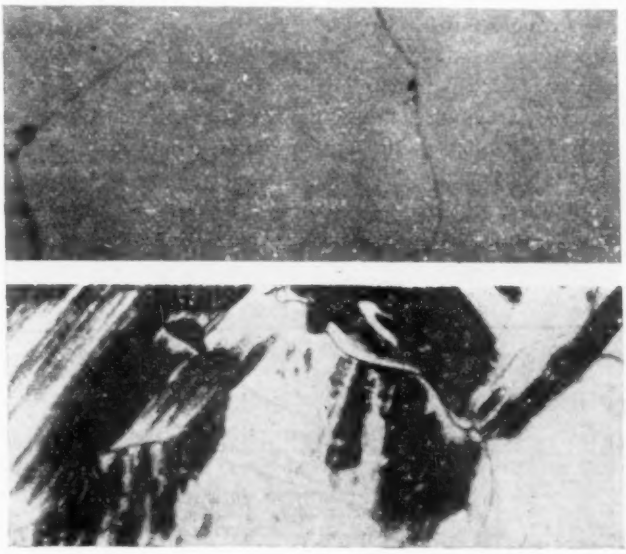


FIGURE 6. *Annealed Armco iron before and after being subjected to 40,000,000 stress reversals of ± 12·2 tons/in² (Gough and Hanson)*

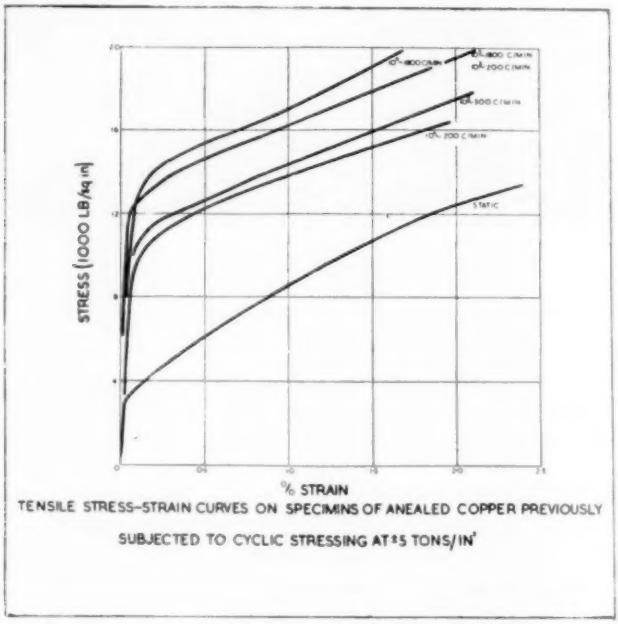
We have now seen the curtain go up on the mechanism of fatigue—it is a melodrama, sometimes a transpontine one, but its start is as gentle as that of a sentimental comedy—the early repeated stresses produce slip bands in certain

Property	Prior to [illegible] reversals	After 10⁷ [illegible] reversals
Ultimate Tensile Stress, tons/in²	42·0	42·0
Reduction of area,	59·1	59·0

FIGURE 7. *Change to static mechanical properties of 0·49% carbon steel as a result of being subjected to 10 million stress reversals at its fatigue limit*

crystals appropriately orientated to the stress, do they do anything else? They do, indeed. The relative movement on atomic planes engenders heat, but the mechanical properties of the metal in which slip bands are developed are also

changed—this is shown in Figure 7, which gives the results of the experiment which was carried out by Moore and Kommers. It shows that repeated stresses which the steel would have withstood indefinitely, have somewhat changed it; it has become stronger but less ductile. That is a very significant fact—repeated stresses destined, as we shall subsequently see, sometimes at long last to create dramatic disaster, initially from many points of view actually strengthen the metal to which they are applied; the influence of these repeated stresses on metal can be compared to that of alcohol on the human mechanism—initially stimulating but, in spite of the maxim 'another little drink won't do us any harm', decreasingly effective and ultimately disastrous. In case this homely analogy is not understood—audiences are always so much more refined than lecturers—



TENSILE STRESS-STRAIN CURVES ON SPECIMINS OF ANEALED COPPER PREVIOUSLY SUBJECTED TO CYCLIC STRESSING AT ±5 TONS/IN²

FIGURE 8

the facts are given in Figure 8. Bullen, Head and Wood subjected testpieces of annealed copper to varying numbers of cyclic stresses of $\pm$ 5 tons/in.^{2}, a stress greatly in excess of the elastic limit of this metal. Subsequently the stress-strain diagrams reproduced in Figure 8 were made. It will be noted how strikingly a hundred cycles have raised the elastic limit of the originally annealed metal. On the other hand, the change in this respect of the specimen subjected to a million cycles, as compared with that to which a hundred thousand were applied, is impressive by its smallness.

What I have shown is evidence as to the nature of a part of the fatigue process. It is characterised by three simultaneous happenings at a decreasing rate:

(*a*) slipband formation,
(*b*) heat evolution, and
(*c*) increasing elastic limit.

If the stress level is relatively low, these three types of change appear to cease. Then for many hundreds of thousands and possibly for millions of further cycles, there is no detectable change in the physical properties of the specimen. Following the stage in which much happens and no damage is done, a new chapter opens, of which one speaks with less confidence. It might be called 'the nothing doing stage', but this is surely the wrong title. It is a stage in which, at present, we can see nothing doing. Generally speaking:

(*a*) no more slip bands are formed;
(*b*) there is no evolution of heat;
(*c*) mechanical properties are unchanged, and
(*d*) the deflection of the specimen under load remains constant.

This stage of mysterious inaction, measured in number of stress cycles may greatly exceed the first one—it is comparable to the calm before the storm.

The outward signs, visible to an engineer, that a third chapter of the fatigue story has opened are these:

(*a*) the deflection under load of the specimen starts to increase;
(*b*) the temperature of the specimen begins to rise;
(*c*) if a stress-strain diagram of the cycle were plotted, there would be a growing hysteresis loop;
(*d*) internal damping would increase.

These happenings are explicable on the basis of one or more cracks having developed in the specimen. The signs, visible to a metallurgist, are these:

(*a*) in a region of high slip-band development, a crack or cracks have come into being which, at least initially, are parallel to a slip band;
(*b*) by means of X-rays it can be shown that in the neighbourhood of the crack there has been a formation of disorientated crystallites.

These happenings culminate in the specimen breaking with a largely brittle fracture, such as was seen in Figures 2 and 3.

Due to the kindness of Dr. N. Thompson of the University of Bristol, I am able in Figure 9 to show a composite picture. In four stages, it tells the story of a piece of high conductivity, fully annealed copper, subjected to repeated stresses. It is not a metallurgical Rake's Progress, but a cautionary tale. It tells of what happens even to very ductile metal, if you treat it rough. Unlike the worm, it does not turn but, more disastrously, cracks.

The first picture shows the surface of a piece of copper which has been electro-polished after it has been subjected to slightly over a quarter of a million stress reversals. Its appearance does not differ from that of virgin metal—no slip bands are to be seen.

The second is of the same specimen under about the same magnification.

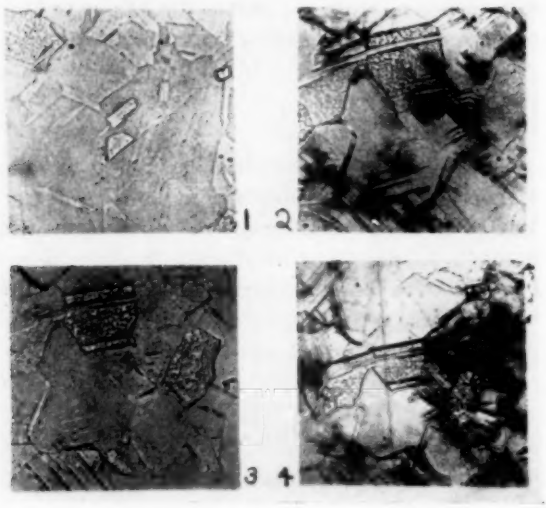


FIGURE 9. *Progressive deterioration of a piece of annealed copper as a result of cyclic stresses*

After about $1\frac{1}{2}$ million stress reversals, slip bands, and perhaps something more, have developed.

The third is of the same portion of the specimen following electro-polishing. This metal-removing process indicates that some bands are more persistent than others; it suggests that they may be of two kinds.

The fourth part of the figure shows the influence of a little over $3\frac{1}{2}$ million stress reversals.

The marks in single crystals which persisted after electro-polishing have extended into other crystals; they are cracks; the end is at hand. In brief, the fatigue story is this:

(*a*) repeated stresses produce slip bands. These increase the strength but reduce the ductility of the metal;

(*b*) further repeated stresses engender a crack at or near a slip band;

(*c*) still more repeated stresses cause the crack to spread beyond the crystal in which it came into being—what was initially a slip has become a landslide.

Figure 9 has shown the metamorphosis of a crack; the next one, for which I am indebted to Mr. Peter Forsyth of the Royal Aircraft Establishment, indicates where a crack might have developed and where, in fact, it did. Figure 10 is a photomicrograph of a metallurgically polished specimen of ALCLAD—BS L38. As a result of cyclic stresses, a number of slip bands have developed. A crack has taken place along one of them. After travelling a certain distance, it changes direction in its almost human search for the line of least resistance.

I have given you a strip cartoon of fatigue failure in what might be called a common or proletarian metal; pennies are, and kettles were, made of copper—the contemporary anæmia of the kitchen is relatively modern. I have followed this by the final picture of fatigue failure in an aluminium alloy which used to move at the highest levels, for it was largely employed in the construction of aircraft.

I now wish to give you another pictorial story, showing the evil influence of cyclic stresses, on a somewhat exotic alloy of aluminium, zinc, magnesium, manganese and copper, currently used for aircraft construction—it is called D.T.D. 683. In Figure 11, for which, and for all information relating to which,

I am indebted to Dr. R. F. Hanstock of High Duty Alloys Ltd., you will see the illustrations, all equally magnified, of the story of this metastable alloy under the influence of cyclic torsion stresses. The first photomicrograph (magnification in the slide is about 700 diametres) shows the original field of a fully heat-treated and so far unstressed specimen of D.T.D. 683—little but an intermetallic compound can be seen. The second picture shows a part of the surface of the same specimen after it has been subjected to a number of reversed torsional stresses; at right angles to the torsional axis, a narrow band of precipitation is seen.

FIGURE 10. *Slipbands and crack in B.S.L. 38 as a result of repeated stresses (Forsyth)*

After further stresses of about the original magnitude, additional bands of precipitation can be noted. Finally, after being further stressed, a crack has developed near to one of the bands of stress-engendered precipitation.

In the case of the annealed copper and in that of the first heat-treated aluminium alloy, cracks have developed on planes having decreased ductility due to work-hardening; these are typical examples of the common or garden mechanism of fatigue. In the case of the complex precipitation hardened metastable aluminium alloy, the mechanism is slightly more complex. Cyclic stresses have produced bands of localized precipitation. These have created weak but more ductile bands contiguous to them; in these zones repeated stresses engender slip, work-hardening and ultimately cracking.

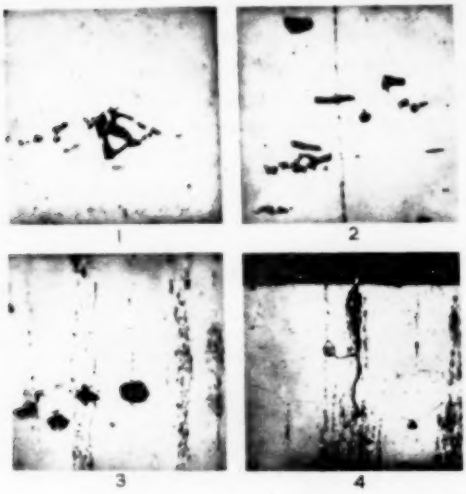


FIGURE 11. *The influence of repeated torsional stresses on the micro-structure of a fully heat-treated specimen of the age-hardened aluminium alloy D.T.D. 683*

PART III. THE RELATIONSHIP OF STRESS TO ENDURANCE

In briefly sketching the mechanism of fatigue, I may have interested some of you—I hope I have—but at this stage, for the sake of the engineers, I must gird up my loins—however this is done—and speak to them in their own language. The engineer, if not distressed by stress is at least obsessed by it—stress is to him what lions are to a lion tamer: his *raison d'être*; he wants to know how many times he can subject a part to a repeated stress without its breaking. The short answer is unfortunately 'try it and try it and try it again, and then don't be too certain of the answer'. This is, indeed, a somewhat unhelpful dictum but I hope to be able to make some useful observations.

More than three-quarters of a century ago, Wöhler applied to nine identical rotating beam specimens, made of axle iron, repeated reversed stresses of different magnitude. Let us look at Wöhler's results, which are set out in Figure 12. They have been gratefully taken from Dr. Gough's Cantor Lecture of a generation ago. You will probably have noted that this is the second time I have quoted from Gough. This makes it appropriate for me to make yet another quotation, not in this case from Gough but from F. G. G. Carr of the National Maritime Museum: 'Copying from one book is clearly cribbing—copying from two or more is research'. On this basis I am hopeful of earning the higher classification.

The thing to note in Figure 12 is how markedly the capacity of axle iron to resist repeated stresses is increased if the applied stress is decreased. Lowering the applied range of stress from $\pm$ 15.3 tons/in.^{2} to 14·3 tons/in.^{2} increases the endurance of the specimen of axle iron from 56,430 to 99,000. This is impressive, but just look at what happens if the stress is approximately halved, that is, reduced from $\pm$ 15·3 tons/in.^{2} to $\pm$ 7·6; the capacity to withstand repeated stresses is increased more than 2,340 times, for the specimen was unbroken after 132,250,000 reversals at $\pm$ 7·6 tons/in.^{2}.

The importance of the fantastic increase in capacity to resist repeated stresses

Specimen No.	Applied Range of Stress. (Tons/in^2).	Number of Repetitions to Fracture.
1	± 15.3	56,430
2	± 14.3	99,000
3	± 13.4	183,145
4	± 12.4	479,490
5	± 11.5	909,840
6	± 10.5	3,632,588
7	± 9.6	4,917,992
8	± 8.6	19,186,791
9	± 7.6	132,250,000 (unbroken)

FIGURE 12. *Wohler's tests on axle. Iron*

which arises from reducing the stress level cannot be too strongly emphasized. This way safety lies, but weight is increased. These are the Scylla and Charybdis of the aircraft designer; at one extreme the aircraft will fail by fatigue; at the other, it will be perfectly safe—it will be too heavy to fly. There is, however, a *via media* which allows of the machine flying for a definite period with definite safety—this involves fantastically realistic testing, the use of large factors of safety and the carrying of an embarrassing amount of weight.

The results of fatigue experiments are not usually given as in Figure 12—graphical presentation is generally employed. A smooth curve is drawn through experimentally derived points showing the relationship of endurance to repeated stress. This form of presentation is called an S/N curve. Figure 13 shows a notable one. It is no ordinary S/N curve. It is, I believe, the best which has yet been made. It is due to T. T. Oberg of the United States Army Air Force Material Command, and relates to the strong wrought American Aluminium Alloy 75S-T6—roughly equivalent to the British D.T.D. 363A.

The curve covers a range of 450 to 8,300 million stress reversals, some made at the rate of ninety per minute, but the majority at 10,600 per minute—even at this speed, the later specimens had to be under continuous test for nearly a year and a half. The number of cycles are plotted on a log scale—the stress range linearly. So plotted, the curve appears to be composed of three distinct parts. On the left-hand side is a steeply sloping, virtually straight line. This indicates that, in this zone, fatigue resistance falls sharply with increase in number of stress

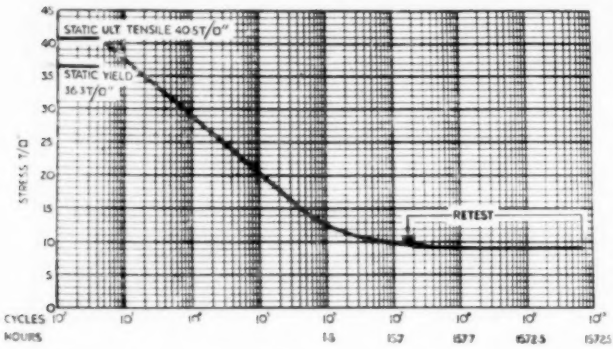


FIGURE 13. *Relationship between cyclic stress and endurance of the aluminium alloy 75 S—T6 (Oberg)*

reversals. On the right-hand side, the curve also approximates to a straight line, but in this case is parallel to the bottom ordinate. This portion of the curve means that, provided the maximum stress does not exceed a certain magnitude—in this case ± 9 tons/in.² —specimens of the alloy in the form in which it was tested would withstand an infinite number of reversals when this is disclosed; the stress at which it occurs is called the fatigue limit of the alloy. Between the two portions of the curve is a transitional one—it is generally called the 'knee'. To the left of it, it would be unsafe to work at the stress ranges shown, except on the basis of a limited number of reversals, to the right, the stress range is safe for any number of reversals. You will notice that if resistance to static tension formed the stress criterion of a design, neglecting any question of a factor of safety, a stress of over 36 tons/in.² could be worked to, but if an infine number of stress reversals are involved, 9 tons/in.² or one quarter of this figure becomes the limiting value.

In Figure 14, one curve relates to hard drawn phosphor bronze having an ultimate tensile strength of 51·5 tons/in.². The other relates to pearlitic carbon steel having an ultimate tensile strength of 37·6 tons in.². Were these alloys under consideration for taking static loads then, leaving out price considerations, probably the phosphor-bronze would be selected. Should repeated stresses have to be resisted, then the phosphor-bronze would still be preferred, unless the number of reversals was in excess of 10^8. You will note that between 10^6 and 10^7 cycles the steel curve becomes parallel to the endurance ordinate—in other words this steel, like the wrought aluminium alloy 75S, has a fatigue limit; if the imposed stress does not exceed ± 14 tons/in.², it will withstand an infinite number of stress reversals. The phosphor bronze curve, even at 10^9 cycles, is still heading downwards; the curve will not, in fact, become parallel to the endurance ordinate; the alloy has no fatigue limit, there is no cyclic stress it will withstand repeatedly—in fact any stress, however small, will break such an alloy, provided always it is repeated a sufficient number of times.

This constitutes a great division in the alloy world—those having a fatigue limit are not numerous, they form a kind of metallic aristocracy and are mostly ferrous; some aluminium alloys possess this desirable quality, but they are a small proportion of the whole. The majority of the alloys, however bravely they first

face the incidence of repeated stresses, wilt in the end—like his late Majesty King Charles II, they may nevertheless be an unconscionable time in dying.

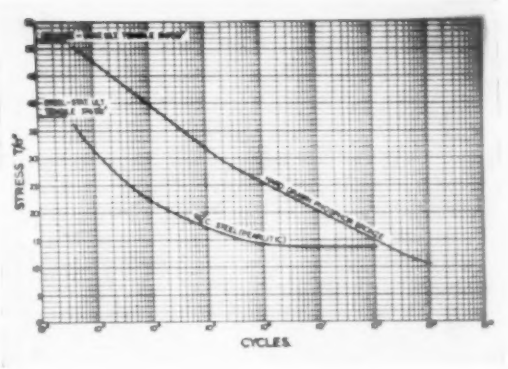


FIGURE 14. *Comparison of the S/N curves of hard-drawn phosphor bronze and of pearlitic carbon steel*

PART IV.
SOME DISTURBING FACTS

With metals, as with individuals, it is their defects and not their virtues which command the greater interest. Rather than give a dissertation on how praise-worthy are the metals to resist repeated stresses, I will draw attention to aspects with regard to which their performance leaves much to be desired.

Scatter

The relationship which I have so far put before you, with regard to repeated stress and endurance at that stress, has been based on the average of a number of experimental results derived from smooth testpieces. Engineers will certainly want to know how greatly individual results, at any particular stress, of such

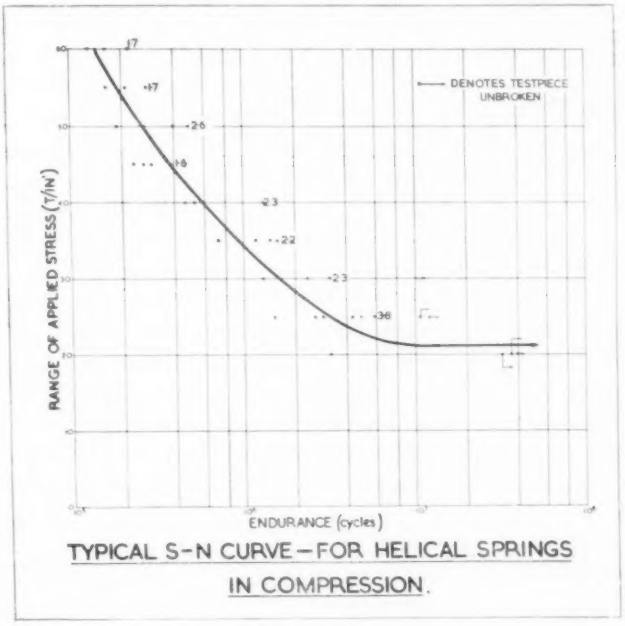

DENOTES TESTPIECE UNBROKEN
RANGE OF APPLIED STRESS (T/IN²)
ENDURANCE (cycles)
TYPICAL S-N CURVE—FOR HELICAL SPRINGS IN COMPRESSION.

FIGURE 15 *(Cox)*

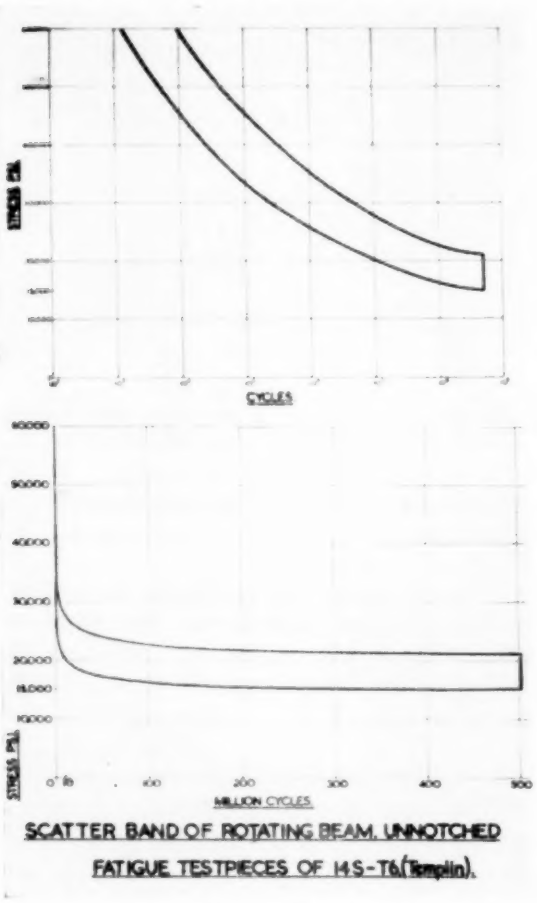


FIGURE 16

specimens from the same batch of metal depart from this average. The answer is certainly a disturbing one—there is always a wide variation in resistance to the same repeated stress of test-pieces, even when made from the same specimen of metal. Variations of comparable magnitude are not observed with regard to static mechanical properties. Thus, once again, it is emphasized how heavy is the burden put upon designers of light structures, subjected to repeated stresses. Figure 15, due to H. L. Cox of the National Physical Laboratory, shows the variation in resistance of helical springs made from the same material and tested under repeated compressive loads. This is an S/N curve on which the points on which it has been based are marked. Under a repeated compressive stress of 60 tons/in.^{2}, the endurance of one specimen was 1·7 times that of another. When, however, the stress was reduced to 24 tons/in.^{2}, the ratio of endurance of high to low was 3·8 to 1 in the specimens which broke—but 'this is nothing', as the Fat Boy in Pickwick might say, wishing to make your flesh creep.

Let us turn to light alloys; Figure 16 tells a disturbing story. The results on which these curves have been based are due in part to the great American experimentalist, R. L. Templin. The plotting has been done in two ways. In both graphs the stress scale is linear; in the top one the endurance cycles are shown on a log scale; in the bottom one, this is linear. Both tell the same story but perhaps with differing degrees of lucidity. Anyway, the curves show the limits of scatter when making an S/N curve for smooth specimens of the wrought Aluminium Alloy 14S–T6. When the maximum stress was 50,000 lb./in.^{2}, the ratio of maximum to minimum was 7·8 to 1 but, when the maximum of the repeated stress was reduced to 21,000 lb./in.^{2} this ration was 48 to 1. The use of

the conventional log scale for number of cycles perhaps makes this difficult to appreciate visually—hence my use of the linear scale. The sole advantages of the log scale are economy in paper and an entirely fictitious suggestion of less scatter in the long endurance results than, in fact, exists.

Before leaving this disturbing aspect of fatigue, namely the variation in endurance of specimens of the same material subjected to the same cycle of repeated stresses, I wish to put before you some very interesting experimental results for which I am indebted to my friend, H. H. Burton of the English Steel Corporation. The experiment involved dividing, on one of the planes of symmetry, a portion of a forged heat-treated alloy steel crankshaft, as disclosed in Figure 17. From one of the halves, tensile and izod testpieces, located as shown in the figure, were cut out and tested. From the other, fatigue testpieces were taken as indicated. On testing the tensile testpieces, the difference in yield point between the lowest and highest was approximately ten per cent. The location of these testpieces is shown in the figure.

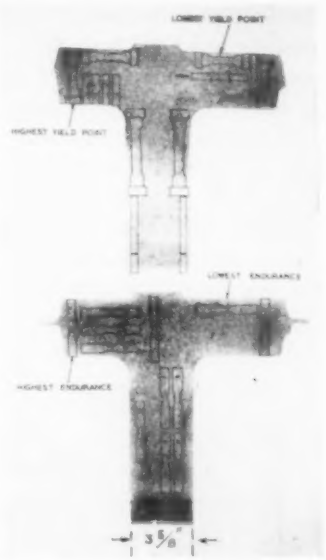


FIGURE 17. *Location of tensile, izod and fatigue test pieces from a portion of a forged heat-treated alloy steel crankshaft*

Turning now to the fatigue testpieces, these were subjected to a cyclic torsion stress approximately 0·4 tons/in.² above the fatigue limit of this steel, when tested in this way. The relative positions of the most and least fatigue resistant testpieces are indicated. They are similar to the tensile ones. Their test results are, however, most impressive—the specimen at location 'K', broke after less than 400,000 stress cycles of the magnitude indicated; that at 'D' was unbroken after almost 15 million cycles of the same stress.

To summarize—the significance of these three quite comprehensive experiments is this:

(*a*) Variation in resistance to repeated stresses of identical specimens from the same batch of material is high in comparison with that in the material's static mechanical properties.

(*b*) The variation in resistance to repeated stresses is greatest when the magnitude of the repeated stresses is low as compared with the material's static ultimate stress.

Stress concentrations

The things engineers design have little or no relation to fatigue testpieces; they invariably contain stress concentrations, such as notches, holes, changes of section, screw threads, and so on. The question therefore arises 'how do stress

concentrations influence fatigue?' An answer is given in Figure 18 relating to a strong wrought-aluminium alloy. In this figure, the vertical ordinate is stress concentration factor. The horizontal one gives the repeated stress which would be endured for 10^7 reversals.

For a stress concentration factor of unity which means there is no notch in the specimen, the alloy under review would, on the average, withstand 10^7 reversals of $\pm$ 14 tons/in.^{2}; for a stress concentration of three which engineers do not very often exceed, the 10^7 endurance has fallen to 8·2 tons/in.^{2}, and when the concentration is forty this figure has become five, which is little in excess of one third of that of a specimen without a notch. The lesson to be learned is this—do everything possible to keep stress concentrations low.

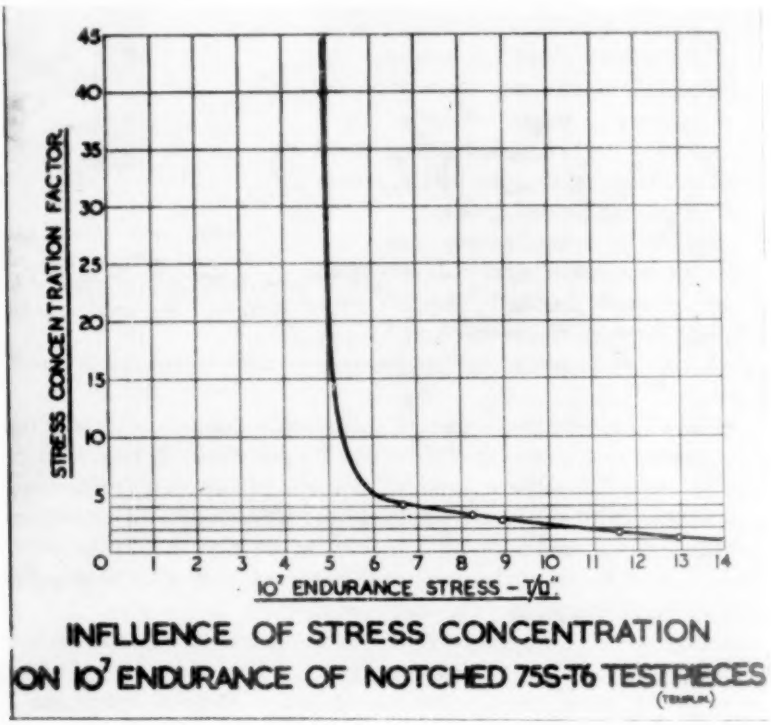


FIGURE 18

If time allowed, similar information could be given with regard to other metals—in general the less ductile an alloy the more adverse is the influence of a stress concentration on its resistance to repeated stresses.

Having indicated how adverse is the influence of a stress concentration, I can elaborate my theme. Dangerous as is the concentration arising from design, which seldom exceeds four, there are other concentrations, some attributable to ignorance and some to accident, which can be far more damaging.

Machining operations, all of which produce local heating, may give rise to hair cracks; this is particularly the case when grinding a metal of low thermal conductivity, such as titanium. Such cracks, because of the stress concentration they create, can have a disastrous influence on fatigue resistance. The list of manufacturing operations which can give rise to stress concentrations could be continued *ad nauseam*—I will merely indulge in a dictum. From the fatigue point of view, it is the surface of the metal which is its Achilles heel, but while the vulnerability of Achilles was limited to a small proportion of his surface, there is no such limitation with a piece of metal—it is all heel and no soul; but then I am no theologian, shoemaker or even fisherman. Throughout its working life, it should be treated as if it possessed the extreme delicacy of a new-born babe, which it indeed does.

Perhaps a word of explanation is desirable here. The inter-crystalline boundary, as has already been said, is almost infinitely thin, but crystals can be of considerable size; in a wrought aluminium alloy one of a tenth of an inch is quite possible. Now stress concentration is a function of the square root of the depth of the crack divided by the radius at the bottom of the crack; so, if by corrosion or other chemical means the almost infinitely thin inter-crystalline boundary is removed for a not very considerable depth—say even for one hundredth of an inch—a fantastically high stress concentration can be produced. It may theoretically easily exceed 1600 and, without anything being visible to the naked eye, the fatigue resistance of a part which has been subjected to this inter-crystalline attack may, in practice, be enormously reduced.

The influence of surface condition on fatigue resistance

The preparation and protection of metallic surfaces so as to have, and to retain, the highest possible fatigue resistance, is a vast subject—time, however, only allows me merely to mention its existence. Nevertheless, it cannot be too strongly emphasized that the importance of a metallic surface subjected to repeated loads arises from the fact that it is at the surface the highest stresses have to be met. In the case of a beam or of a shaft, engineers will readily accept this. With the truly axially loaded tie or strut, they may reject my dictum. It should, nevertheless, be accepted. On the exposed surface of a piece of metal there are exposed crystal faces. Here, because of the elastic anisotropy of crystals and their random orientation, local stresses both higher and lower than the mean cross-sectional one will be engendered.

Since surface stress is a condition antecedent to fatigue failure, it will be at once appreciated that anything which will reduce such stress will tend to increase fatigue resistance. At this stage, I have to ask you to accept from me that repeated loads creating surface tension stresses are more damaging than loads giving rise to compression ones of the same magnitude. This is a considerable pronouncement, for it gives a hint of how the demon fatigue can, at least, be held at bay. By a variety of mechanical or metallurgical processes, it is often possible to put the surface of a metal part into a permanent state of static compression. This has a most excellent influence on its fatigue resistance, if it is subjected to

any form of repeated loading giving rise to tension stresses, for the resultant instantaneous stress acting on the surface of the part will be the algebraic sum of the original compressive stress and of the applied tension one.

As my time is running out I am going to indicate how, by the collaboration of technologists, the danger of fatigue failure can be most markedly reduced. The designer can do his bit; he can so select his proportions that the maximum stress is nowhere high; he can and should avoid serious stress concentrations. The production engineer can also give a most useful contribution; he can make parts with a good surface finish, free alike from inter-crystalline or other cracks produced by damaging surface treatments and above all, free from residual

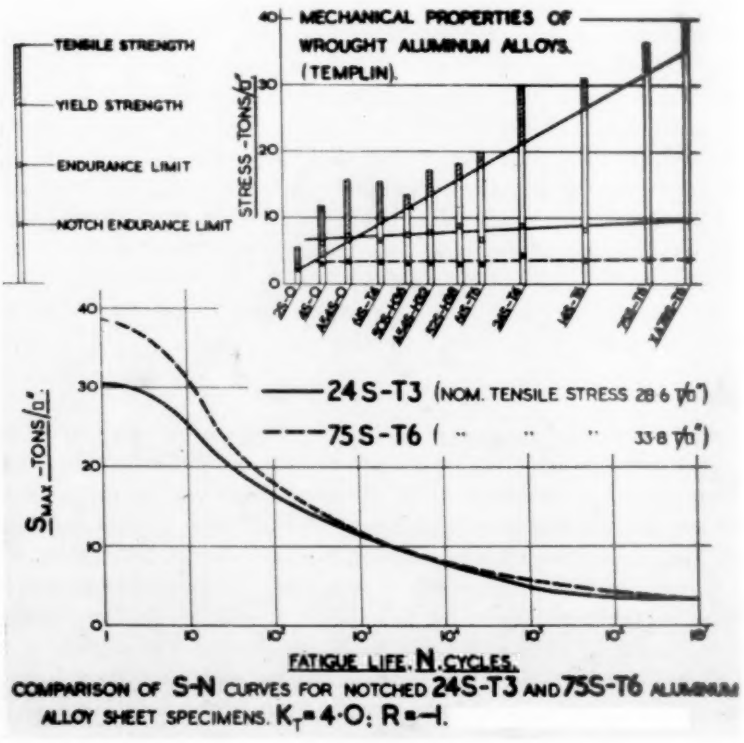

TENSILE STRENGTH
YIELD STRENGTH
ENDURANCE LIMIT
NOTCH ENDURANCE LIMIT
MECHANICAL PROPERTIES OF WROUGHT ALUMINUM ALLOYS. (TEMPLIN).
STRESS - TONS/□"
24 S-T3 (NOM. TENSILE STRESS 28·6 T/□")
75 S-T6 (" " " 33·8 T/□")
S MAX - TONS/□"
FATIGUE LIFE, N, CYCLES.
COMPARISON OF S-N CURVES FOR NOTCHED 24S-T3 AND 75S-T6 ALUMINUM ALLOY SHEET SPECIMENS. K_T = 4·0; R = -1.

FIGURE 19. (*N.A.C.A. tech. note 332*)

tension stresses. The metallurgist, by prescribing appropriate mechanical or metallurgical treatments, or both, and their sequence, can often provide the surface of a part with those qualities which will secure for it the highest resistance to fatigue.

To speak in somewhat Euclidian terms, but with a deep conviction of their truth, it has not as yet been sufficiently realized that components made in conformity to the *same* drawings, from material to the *same* specification do not

necessarily have the *same* capacity to withstand the *same* number of repeated stresses of the *same* magnitude; the nature and sequence of manufacturing operations is vital.

The aluminium alloys

My disturbing facts have so far been of a generalized character. I now have to draw your attention to something else which is more specific and is certainly disappointing. It is impossible to state in dogmatic terms that fatigue resistance of metals bears any close relationship to any static mechanical properties they possess. As a general rule, although as with many if not with all such rules there are the inevitable exceptions, it can, however, be said that in many alloy groups fatigue resistance is proportionally higher, the higher the static ultimate tensile of the alloy.

The aluminium alloy group is disappointing in this respect. In Figure 19 there is confirmation of this. In the top right-hand corner of the figure, a number of American aluminium alloys, most of which have British counterparts, have been set out so that a straight line passes through their static yield points. So spaced, it will be noted that the line passing through their respective endurance limits rises from left to right; that is to say, the fatigue resistance of smooth specimens does somewhat increase with their respective static yield strengths. Now for what is disappointing; the line passing through the notch endurance limit of the different alloys shows, from left to right, virtually no rising tendency. To the designer of the structure of aircraft intended to have a long life, this must be disappointing indeed.

I hoped to be able to say that while this diagram shows that on the basis of some 500 million stress reversals the notch fatigue—with a notch giving a stress concentration of about forty—of one aluminium alloy is pretty much like another, but for a *limited* number of repeated stresses, there was a substantially better resistance on the part of the stronger alloys. While not exactly a giver of cold comfort, such comfort as I can present on the basis of limited experience is not very hot. My evidence is shown in the lower portion of the figure. This indicates that, with two aluminium aircraft alloys differing in static tensile stress by as much as 5·2 tons/in.^{2}, for more than one thousand stress reversals, the endurance of notched specimens having a stress concentration of four is the same.

PART V. SUMMARY AND CONCLUSION

Let me now try to summarize what, in this wholly inadequate sketch, I have tried to put before you:

(*a*) The fatigue of metals is no new discovery.

(*b*) Fatigue failure of even ductile metals is characterized by a largely brittle fracture.

(*c*) With repeated stresses of a magnitude normal in engineering practice, fatigue takes place in three stages. The first is characterized by:

(1) slipband formation;

(2) heat evolution;

(3) increasing elastic and tensile strength, combined with decreasing ductility.

The feature of the second stage is:

(1) its relatively long duration, and

(2) absence of detectable physical changes.

The final stage is normally of short duration. It involves:

(1) the formation and development of a crack or cracks at or near a slip band;

(2) increase in deflection under load;

(3) evolution of heat, and

(4) increase in damping capacity.

(*d*) The ability of a piece of metal to withstand repeated stress is related to the magnitude of those stresses—endurance increases with decrease in the magnitude of the imposed stress.

(*e*) With certain metals, if the stress is decreased sufficiently, endurance to such cyclic stresses apparently becomes infinite.

(*f*) With other metals, any repeated stress, however small, will produce fatigue failure should it be repeated a sufficient number of times.

(*g*) There is very great variation in the capacity of specimens of the same material having virtually identical static mechanical properties to withstand the same repeated stress.

(*h*) The surface condition of a piece of metal has a profound influence on its fatigue resistance.

(*i*) While in the last generation there have been very great increases in the static mechanical properties of the wrought aluminium alloys, there have been no corresponding increases in their fatigue resistance, particularly where high stress concentrations are concerned.

Conclusion

I have not told even all I know about the fatigue of metals, but all I can tell you in the time available. Of course, there is far far more which is known, but there is also much which is still unknown. To me, there are but two ways whereby the unknown can become known: as a result of well-planned research, or as a result of hurried inquiries engendered by disaster; there can be no doubt as to which is the right approach and to emphasize this, I will quote some words uttered, over two centuries ago, by Stephen Hales, a Fellow of the Royal Society:

> And since we are assured that the all-wise Creator has observed the most exact proportions of number, weight, and measure in the make of all things, the most likely way, therefore, to get any insight into the nature of those parts of the creation which come within our observation must in all reason be to number, weigh and measure.

GENERAL NOTES

LONDON EXHIBITIONS

Following a series of exceptionally interesting one-man shows in the late spring, enterprise in the London galleries now shows signs of slackening a little, though artistic activity may be expected to continue until August when dealers, having hung their mixed shows, customarily pack their bags and depart in search of fresh talent or any masterpieces still unearthed on the Continent. An event of unusual importance in the art world will ensure, however, a good deal of activity in the London galleries early in July, when foreign art critics are to be entertained here by the British section of the *Association Internationale des Critiques d'Art*, an influential body associated with U.N.E.S.C.O. that has its central office in Paris.

Last year the Association's fifth international congress was held in Istanbul. This time the delegates, headed by M. Paul Fierens of Belgium, will occupy rooms at Wadham College, Oxford, and be officially welcomed in the Taylorian by the Vice-Chancellor, and the President of the British Section, Mr. Eric Newton. Lectures in the Taylorian, including the inaugural one by Professor Boase, and work sessions with various aspects of art and the education of the public in these uneasy times, will occupy the critics until 6th July, when the gathering is transferred to London. Thereafter a Government reception, another arranged at the Tate in rooms specially hung with the pick of the British School in the Gallery, with visits to Sir Kenneth Clark's collection at Saltwood Castle, to the Wildenstein Gallery's important exhibition of seventeenth-century Roman painting, and to the studios of our most prominent artists, should leave the distinguished guests, it is to be hoped, with favourable impressions of our enterprise and standing in the world of art, and of our reciprocal hospitality.

No special visit has been arranged for the delegates to Kenwood, though the grounds are always a joy, and the mansion at present houses a pleasing loan collection of paintings by Angelica Kauffmann which will remain on view until the end of September. This artist, who was born in Switzerland and came to England in 1766, staying here until 1781, is no doubt best remembered as an original member of the Royal Academy, the charm of her painted ceilings being still apparent in the Academy's entrance hall. That she was also a skilful portraitist is seen, for example, in her self-portrait in the National Portrait Gallery; but her elegance could be somewhat vapid as well as mannered, and indeed much of the attraction of the Kenwood exhibition derives rather from the light it throws on 'history painting' on classical themes and the early aims of the Royal Academy.

Meanwhile, one of the most discussed painters of the School of Paris has come to Tooth's Gallery, Bruton Street, in the person of M. Antoni Clavé. Born in 1913 in Barcelona, where he later attended the School of Art, Clavé did not come into prominence until some years after his arrival in Paris when he was commissioned to design the sets for the ballet *Los Caprichos* in 1946. This was followed by his *décor* at various times for Sadler's Wells and Roland Petit, and for the film of *Hans Christian Andersen*—so that the success of his last show of paintings at the Galerie Drouant-David was already assured, as indeed was the current Tooth's exhibition on the opening day.

Antoni Clavé's subjects, mostly studio 'lay-figures' given a semblance of life, culinary objects and landscape, provide the departure-points for conceits that are very much of the School of Paris, yet also disclose a Spanish strain. No doubt his Catalan origin would be more apparent could we have seen his religious paintings, though it is suggested in his *Cavalier* and in the emotional overtones of other figure paintings in sombre yet luminous tones. Picasso's is among the influences of this eclectic artist whose rich confections should not be overrated. But the point to

stress is that the best of them are capable of haunting the imagination, as well as exciting the eye by inventive orchestrations of colour and a supremely confident manipulation of paint which we have come to expect of Parisian favourites.

It is another romantic artist, who can likewise claim some notable sets for ballet and opera, who returns to the Leicester Galleries with a display of virtuosity which, one must own, begins to pall a little. It is not just that Mr. John Piper's landscapes—or, more exactly, settings—seem to demand figures and possibly stage movement to make them convincing, for one never required these in his glowing accounts of bombed buildings. But, in wartime, Mr. Piper's sense of drama exactly suited his uncanny theme. To-day he invests with a theatrical air, and subjects to the same tricks of texture and fluid scribble, every peaceable scene he sees. Precisely the same mood is conveyed in lodge gates, lighthouses, and so on, all round the room, apart from some lively variations on ancient seals which make an immediate effect. The truth seems to be that this fertile artist has become so accustomed to devising his brilliant effects for the stage, that he has lost his sense of the distinction between the artificial splendours properly enclosed by the *proscenium* arch and the art of the easel picture which requires a more prolonged and undistracted scrutiny.

NEVILE WALLIS

BRENTFORD 'DUO-THERM' HEAT PUMP

An interesting application of the principles of the heat pump to domestic purposes is provided by the Duo-Therm heat pump, which was recently included in the 'Research in Industry' display at the Castle Bromwich section of the B.I.F. This device provides at the same time a source of heat and cold for the household. The Duo-Therm if installed in a pantry will provide a refrigerator, keep the whole pantry at a temperature of 40 degrees fahrenheit, and keep the water in a 20-gallon hot-water tank at 140 degrees fahrenheit, providing 120 gallons of hot water a day. At the same time a warmer, run off the Duo-Therm, may be fitted in the airing cupboard. The running costs of these four services are only half that of electric water heating alone. The Duo-Therm needs only to be connected with the piping system of the house, and with the water storage tank in the roof. It plugs into an ordinary three-pin lighting point.

ANTOINETTE MONOPLANE

The 'Antoinette 1910 Model' monoplane, which has been on loan to the Science Museum since 1926, has been presented for permanent inclusion in the National Aeronautical Collections. The aircraft, of French design and manufacture, was brought to this country in 1910 by Hubert Latham who had previously attempted to cross the Channel with earlier types of this machine. It is the only complete one of its make and type now in existence, and was recognized in its time as being capable of flying in bad weather. The 'Antoinette', together with the 'Bleriot', established the now conventional form of monoplane with tractor air screw and tail plane.

ULSTER FOLK MUSEUM

Acting on the recommendation of the Nugent Committee, 1954, the Government of Northern Ireland has provisionally undertaken to help finance a Folk Museum illustrative of Ulster provincial life of the past, and of traditional arts and crafts. The Museum, once established, is intended to be an educational institution and at the same time a centre where ancient handicrafts and early machine processes may be fostered. The success of the scheme depends upon support being forthcoming not only from the local authorities in Northern Ireland, but also from the general public. An illustrated booklet describing the project can be obtained free from: The Director, Belfast Museum and Art Gallery, Stranmillis, Belfast, Northern Ireland.

CARPET COMPETITION

Furnishing announces its seventeenth Carpet Design Competition. Prizes are offered by the organizers and also by the Federation of British Carpet Manufacturers and by the Orient Steam Navigation Co., Ltd. The Competition is open to students and everyone in the carpet trade. The closing date for the receipt of designs is 19th September, 1955. Full particulars may be obtained from: The Editor, *Furnishing*, Drury House, Russell Street, London, W.C.2.

NOTES ON BOOKS

GEORGE STEPHENSON; CREATOR OF BRITAIN'S RAILWAYS. *By John Rowland. Odhams Press*, 15*s*

While the purist may object that the sub-title of this book, 'Creator of Britain's Railways', is less than just to many others with vision who had a share in this creative work, there can be no doubt that George Stephenson was the dominant figure. The fact is the more remarkable in that young Stephenson had no advantage of birth or education; nothing but dogged determination and sheer force of character could have carried him through to the eminence of his later years, possessor of a name known the world over. The author has done full justice to this epic story.

So far from receiving a large measure of public support for his ideas, Stephenson had to fight the most severe opposition, especially when he was called in to act as principal engineer to the Liverpool and Manchester Railway. At the preliminary inquiry, even the lawyer who represented the promoters warned Stephenson strongly against suggesting the possibility of so high a speed as twenty miles an hour over the railway; he would be regarded as 'a maniac out of Bedlam' if he did so. The lawyer on the other side did his utmost to discredit this simple witness, but with the cleverest of his questions found it difficult to confound the cool-headed and quick-witted Northerner.

Caste also weighed heavily against Stephenson in these early days. Engineers who had the advantage of means and such technical training as was then available were loath to admit the self-taught Stephenson, however brilliant he might be, on a basis of equality. One of the bitterest pills that he had to swallow was when the committee which had been formed in 1815 to improve the safety of working in mines awarded £2,000 to the renowned Sir Humphry Davy for his safety lamp, whereas Stephenson, who simultaneously submitted an equally efficient lamp, received a beggarly £100. It says a good deal for Stephenson's reputation at that time that he was later presented with £1,000, raised by public subscription, in recognition of his invention.

This is a reminder of the wide range of Stephenson's activities. From the moment when, at the modest remuneration of a shilling a day and at the age of fourteen he was appointed assistant engineman at Black Callerton Colliery on Tyneside, his passion was to take all kinds of machinery to pieces and put it together again, often with material improvements, with the result that in his early twenties his reputation as an 'engine doctor' was known far and wide. The same principle he applied to watches and clocks, until he was supplementing his meagre income on a substantial scale by watch and clock repairing.

It was Stephenson's appointment as enginewright at the Killingworth High Pit that proved the turning-point in his career, and here he built his first steam locomotive with the roughest of tools. Even so, his work was so ignored by engineers and others in a position to influence great enterprises that he was within an ace of packing up and crossing the Atlantic to seek his fortunes in the New World. Had he done so, railway history might well have been written differently. But then followed his association with Edward Pease and the opening of the Stockton and Darlington

Railway, the engineering and equipment of which had been his own work throughout and, finally his appointment four years later as principal engineer of the Liverpool and Manchester Railway at an annual salary of £1,000. With these two railways, the first steam-operated lines in the world to be opened for public traffic, his name and fame had been made; from then on he never looked back.

In these days when equality of opportunity and maximum reward for minimum effort are the aim of so many, this deeply interesting story of how one man, at least, realized that 'blood and tears, toil and sweat', allied to an unswerving purpose, can bring the most rewarding success of all, should prove an inspiration to any young man to-day who is not afraid of hard work.

CECIL J. ALLEN

SHORT NOTES ON OTHER BOOKS

GROWING UNDER GLASS. *Edited by F. A. Mercer and Roy Hay. Studio*, 1955. 15*s*

This sixth volume of 'Gardens and Gardening' contains chapters by authorities on various aspects of growing under glass. The section on heating gives much advice on latest developments. A survey of the newest labour-saving garden aids is included at the end of the book. There are many photographic illustrations.

FOUNDATIONS IN THE DUST. *By Seton Lloyd. Penguin Books*, 1955. 2*s* 6*d*

The story of archæological exploration in Mesopotamia is given, from the discoveries of the great pioneer Assyriologists of the last century down to the present day. The lives and accomplishments of the early archæologists are combined in a single narrative, and a picture is presented of many great personalities. A map is included.

FROM THE JOURNAL OF 1855

VOLUME III. 8*th June*, 1855

From the Report of the eighteenth anniversary meeting of the Yorkshire Union of Mechanics' Institutes.

On the subject of female education, the report says: 'The reports and facts furnished by the Institutes show that, in two or three instances, increased attention is being paid to Female instruction. For instance, at Rotherham there are fifty females receiving instruction. At Bradford there are the same number. At Stanningley the Female classes succeed. There is, however, we regret to say, no proportionate progress made in Female instruction compared with its real importance. Everyone asserts the necessity of Female education, that it is more important than that of the other sex,—that the education of woman is the education of both man and woman begun in the right place, and other generalities are from time to time uttered on this suggestive topic. But as an admitted principle, demanding practical realisation without any further delay, it seems to meet with far too small recognition. The most noteworthy exception to this remark is in connection with the Leeds Institution, where a Girls' Day School, of very high character, is being established. The course of instruction, besides the ordinary accomplishments of the usual boarding-schools, comprise departments which *all* schools *ought* to embrace, and which will certainly confer upon the pupils advantages of no ordinary value, such as Natural History, Natural Philosophy, Elementary Chemistry, and Physiology'. At the Churwell, Headingley, and Northowram Institutes Sewing Classes have been established, one person of superior education being employed to read aloud to the rest while at work.